Transactions
AMERICAN FOUNDRYMEN'S ASSOCIATION

East Engin. Library
NOV 24 1941

Contents

Effect of Pouring Temperature on the Strength and Microstructure of Gray Cast Iron, *by G. A. Timmons and V. A. Crosby* 225

The Mechanism of Pin-Hole Formation, *by C. E. Sims and C. A. Zapffe* 255

Classification of Foundry Cost Factors — Report of A.F.A. Cost Committee 282

A University Course in Foundry Control Methods, *by Fulton Holtby and H. F. Scobie* 310

Some Causes for Porosity and Leakage in Non-Ferrous Castings, *by A. H. Hesse* 331

Heat Treatment of Cast Iron, *by C. A. Nagler and Dr. R. L. Dowdell* 361

Some Properties of Synthetically Bonded Steel Molding Sands, *by R. E. Morey and H. F. Taylor* 388

Report of Steel Division Committee on Radiography—1941 ... 427

The Velocity of Conversion of Austenite to Ferrite and Cementite, *by H. A. Schwartz and M. K. Barnett* 441

VOL. 49 DECEMBER, 1941 NO. 2

TRANSACTIONS

AMERICAN FOUNDRYMEN'S ASSOCIATION
INCORPORATED

Published and copyrighted by the American Foundrymen's Association, Inc., 222 West Adams Street, Chicago, Illinois

Entered as second class matter September 9, 1938 at the postoffice at Chicago, Illinois, under the Act of March 3, 1879

Issued Quarterly

Subscription Price:
Members:
$4.00 per year; $1.00 per copy

Subscription Price:
Non-Members:
$15.00 per year; $4.00 per copy

VOL. 49 DECEMBER, 1941 NO. 2

A.F.A. Organization and Personnel — 1941-42

Officers and Staff

**President*, H. S. Simpson, National Engineering Co., Chicago, Ill.

**Vice President*, D. P. Forbes, Gunite Foundries Corp., Rockford, Ill.

**Executive Vice President and Treasurer*, C. E. Westover[1].

Convention and Exhibits Manager, C. E. Hoyt[1].

Secretary, R. E. Kennedy[1].

Assistant Secretary, N. F. Hindle[1].

Assistant Treasurer, J. Reininga[1].

Board of Directors

Terms Expire 1942

W. B. Coleman, W. B. Coleman & Co., Philadelphia, Pa.

C. R. Culling, Carondelet Foundry Co., St. Louis, Mo.

Otto A. Pfaff, American Foundry Equipment Co., Mishawaka, Ind.

Marshall Post, Birdsboro Steel Foundry & Machine Co., Birdsboro, Pa.

*Fred J. Walls, International Nickel Co., Detroit, Mich.

Terms Expire 1943

*G. W. Cannon, Campbell Wyant & Cannon Foundry Co., Muskegon, Mich.

B. D. Claffey, General Malleable Corp., Waukesha, Wis.

Harold J. Roast, Canadian Bronze Co., Ltd., Montreal, Canada.

L. P. Robinson, Werner G. Smith Co., Cleveland, Ohio.

*H. S. Washburn, Plainville Casting Co., Plainville, Conn.

Terms Expire 1944

R. J. Allen, Worthington Pump & Machinery Corp., Harrison, N. J.

J. G. Coffman, Los Angeles Steel Casting Co., Los Angeles, Calif.

W. J. Corbett, Atlas Steel Casting Co., Buffalo, N. Y.

M. J. Gregory, Caterpillar Tractor Co., Peoria, Ill.

*L. N. Shannon, Stockham Pipe Fittings Co., Birmingham, Ala.

* Member Executive Committee.
[1] Association Office, 222 West Adams St., Chicago, Ill.

Effect of Pouring Temperature on the Strength and Microstructure of Gray Cast Iron

By G. A. Timmons* and V. A. Crosby*, Detroit, Mich.

Abstract

Experiments are described which indicate that a reduction of the pouring temperature reduces the strength of irons of the high test class but produces relatively little changes in the tensile properties of lower strength irons. The lower the pouring temperature, the greater is the tendency to form graphite distributions of the whorl, dendritic and eutectiform types and the greater is the quantity of ferrite found in the microstructures of test bars. Changes in microstructures are correlated with changes in tensile properties as the pouring temperature is reduced. The similarity of ferrite in gray iron, ferrite in chilled iron, and the microstructure of box cooled, hypereutectoid cases of carburized abnormal steels is discussed.

1. It has been known for some time that cast iron undergoes a change in characteristics when the pouring temperature is reduced. To avoid misinterpretation, it must be understood that throughout this paper the term "pouring temperature" applies to the temperature of the molten iron as it enters the mold. Since most jobbing foundries usually receive the metal from the cupola in a large bull ladle, from which smaller ladles are filled and the castings are poured from these smaller ladles, an appreciable drop in temperature occurs during the pouring of a single run. Castings poured during the early part of the period will be cast at considerably higher temperatures than those poured during the latter part. Casting of most electric furnace heats is done with small hand ladles, often requiring 5 to 10 min. to cast the entire heat and, unless power is delivered to the furnace during the pouring of the heat, there is a significant drop in temperature between the first and last castings poured. Since this practice prevails in most foundries, it should be to the advantage of the industry to know

* Metallurgist, Climax Molybdenum Co.

Note: This paper was presented before a Gray Iron session during the 45th Annual A.F.A. Convention, New York City, N. Y., May 15, 1941.

what differences in specific properties can be expected from the variations in pouring temperature. It is the purpose of this publication to present quantitatively the effects of reductions in pouring temperature upon the tensile strength of various cast irons and to correlate the changes in tensile strength with changes in microstructure.

REVIEW OF THE LITERATURE

2. DiGiulio and White[1] have studied the effects of variations in superheating and pouring temperatures on gray cast iron and, with regard to pouring temperature, showed that "an optimum pouring temperature exists which varies for different irons . . ." and that irons poured above and below this temperature show poorer tensile properties than bars cast at this temperature. Unalloyed irons were used for these tests and strengths ranged from 43,000 to 31,000 lb. per sq. in.

3. Bolton[2] has shown the effect of pouring temperature on the tensile strength of cast iron. He cites an example using the following composition:

Carbon, per cent	2.74
Silicon, per cent	2.25
Phosphorus, per cent	0.22
Manganese, per cent	0.52
Sulphur, per cent	0.056

When this composition was cast into 1.2-in. diameter test bars with the metal at a temperature of 2620°F., the tensile strength was found to be 38,900 lb. per sq. in. When the pouring temperature was reduced to 2335°F., the tensile strength of similar test bars was 31,550 lb. per sq. in. An examination of the microstructures of the two test bars revealed that the bar cast at the higher temperature consisted of pearlite with a random distribution of graphite flakes, while the iron poured at a lower temperature consisted of flakes arranged in whorl or rosette patterns and the latter iron contained appreciable quantities of ferrite.

4. Roth[3] has stated that high strength irons should be poured as hot as possible to obtain maximum properties. To illustrate this point, he indicated that over a range of temperatures from 2720 to 2575°F., the strengths of the alloyed irons were 60,000 lb. per

[1] Superior numbers refer to bibliography at the end of this paper.

sq. in. or above. However, when the metal was allowed to cool to 2475°F. before casting, the strength was found to have dropped to 49,640 lb. per sq. in.

5. Table 1, taken from Alloys of Iron and Carbon[4] Vol. II, indicates the effect of pouring temperature upon the tensile strength of cast iron.

Table 1

EFFECT OF POURING TEMPERATURES ON THE TENSILE STRENGTH OF CAST IRON[4]

Casting Temperature, °F.	*Tensile Strength, lb. per sq. in.*
2515	20,900
2445	23,600
2380	22,900
2320	23,000
2245	24,300
2155	23,500

6. Ziegler and Northrup[5], in discussing the effect of superheat on cast irons, made the following observation after examining test bars of practically the same composition poured at different temperatures with variations in superheating temperature:

> "Invariably, in each case, the sample of higher tensile strength has more pearlite and less ferrite than the sample of lower tensile strength. Moreover, in the case of the former, graphite flakes are more 'localized' and separate from each other than in the case of the latter. In the latter case, graphite flakes tend to form more or less continuous paths sometimes associated with formation of dendritic pattern."

> Six photomicrographs were presented in three pairs comparing the microstructure of the test bar exhibiting the highest tensile strength with that of the test bar exhibiting the lowest tensile strength in each group. In every case, the sample exhibiting the lower tensile strength was poured at the lower temperature. Table 2 is taken from their work.

7. While it is recognized that superheating plays an important role in determining microstructures and physical properties of an iron, it must also be recognized that the pouring temperature of iron CP was 310°F. below that of BS; the pouring

Table 2

EFFECT OF POURING TEMPERATURE ON PHYSICAL PROPERTIES OF CAST IRONS[5]

Sample No.	*BL*	*BQ*	*BS*	*CP*	*CY*	*DO*
Chemical Analyses, per cent						
Total Carbon	2.88	3.00	2.46	2.58	2.18	2.24
Silicon	2.31	2.15	2.29	2.25	2.29	2.29
Manganese	0.53	0.47	0.46	0.45	0.41	0.40
Sulphur	0.048	0.088	0.068	0.072	0.044	0.054
Phosphorus	0.38	0.39	0.36	0.36	0.32	0.30
Superheating Temp., °F.	2800	2600	3000	2600	3000	2600
Pouring Temp. (Measured), °F.	2600	2430	2870	2560	2720	2560
B. H. N.	217	207	229	229	241	241
Tensile Strength, lb. per sq. in.	42,350	31,100	55,000	36,000	57,300	34,500

temperature of iron BQ was 170°F. below that of BL; and the pouring temperature of iron DO was 160°F. below that of CY.

8. These references, pertaining to a relationship between tensile strength and pouring temperature, and some experiences in the laboratory production of experimental cast iron heats, led the authors to conduct an investigation to ascertain the importance of pouring temperature on both alloyed and unalloyed irons, with a consideration of the accompanying microscopic changes to gain some knowledge of the reasons for the observed relationship.

EXPERIMENTAL PROCEDURE

9. Eight separate heats were studied and four A.S.T.M. (A48-36) *B* test bars were cast from each heat. These heats were melted down in a 30-lb., silica-lined, high frequency induction furnace. With the exception of heats 1811 and 1853, the raw materials used for the production of the experimental heats consisted of Armco ingot iron punchings, Acheson graphite granules, ferrosilicon, ferromanganese, ferrochromium, ferrophosphorus, ferrous sulphide, copper scrap and XX nickel shot. Heat 1811 was melted down with approximately 50 per cent chateguay pig iron in the charge while the charge for heat 1853 contained 75 per cent of the same pig iron. The remaining raw materials for these heats were the same as for the others. Late additions of silicon were made to heats 1810, 1811, 1853 and 1867. In these heats 0.30 per cent silicon (as 75 per cent grade ferrosilicon) was kept out of the charge, which was completely melted and superheated to 2800°F. before the late silicon addition was made.

10. In the case of the other four heats, all of the raw material was melted down in the charge. In all cases the melt was superheated to 2800°F., the highest temperature generally used in cupola practice. When a late addition of silicon was added, the metal was held 2 min. to ensure complete solution of the late addition and uniform distribution throughout the heat. The first test bar was then poured. The metal was cooled to 2650°F. and a second test bar was poured. The third and fourth test bars were cast at 2500°F. and 2350°F. respectively. Temperatures were measured by an optical pyrometer, applying a correction for the surface of the bath in the open.

11. Test bars were cast in baked sand molds with their axes in a vertical direction. All bars were cooled to room temperature in the molds.

12. Impact test specimens, 1.125-in. diameter by 8-in. long, were machined from the lower halves of the test bars. These were broken in the Charpy impact testing machine. A *B*-tension test specimen was machined from each broken impact test bar. After the tension specimen had been broken, a metallographic specimen was prepared from a section in the gauge length. The metallographic specimens were examined at various magnifications.

13. Photomicrographs were prepared to illustrate the changes effected by a reduction in pouring temperature. With the exception of heats 1867 and 2280, the irons contained pearlitic matrices with varying quantities of ferrite. Results of the tests are given in Table 3 and photomicrographs are shown in Figs. 1 to 4.

SUMMARY OF DATA

14. A study of the data in Table 3 indicates that pouring temperature is not an important factor in determining the tensile strength of unalloyed or low strength irons, even though the microstructure is altered perceptibly by a reduction in the temperature at which the metal is cast. In these irons, typified by heats 1810, 1811 and 1853, the refinement in the graphite flakes apparently offsets the effect of an increase in ferrite.

15. In the higher strength irons, pouring temperature plays an important role in governing the tensile strength of an iron. Here the presence of graphite distributions characteristic of under-

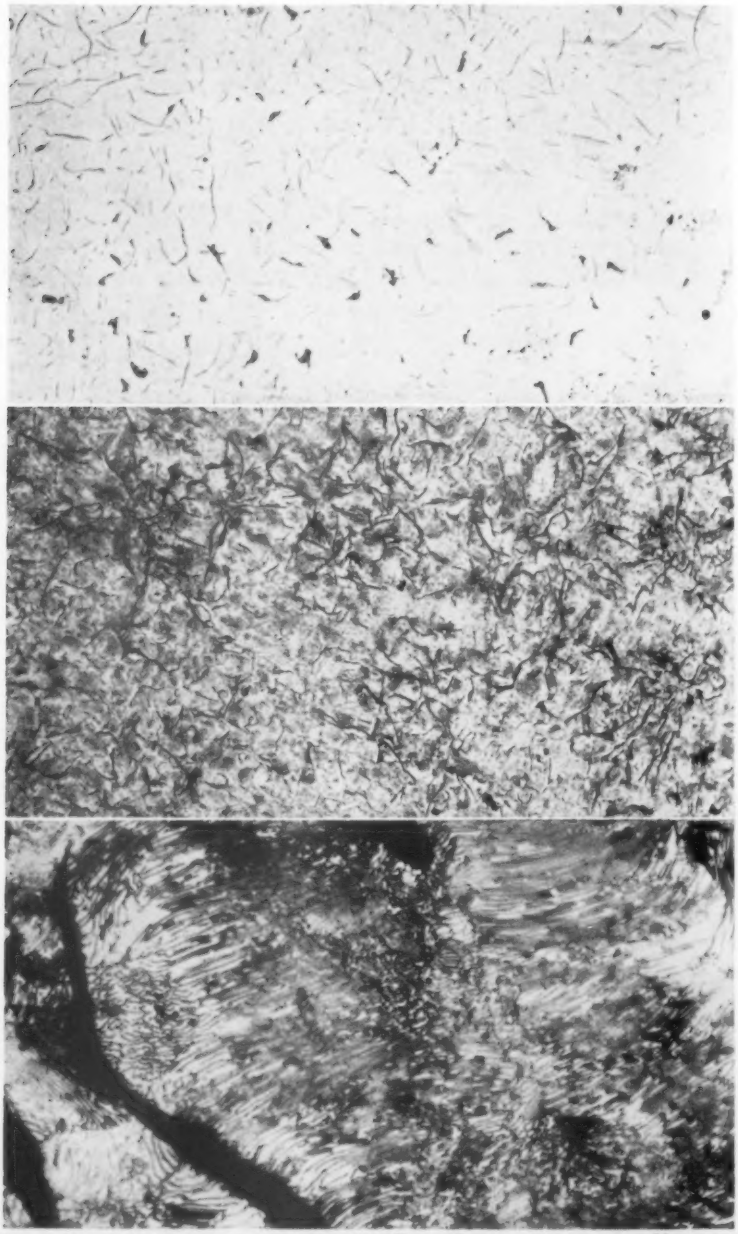


FIG. 1—HEAT 2277 CONTAINING 3.00 PER CENT TOTAL CARBON, 0.75 PER CENT MANGANESE, 2.00 PER CENT SILICON, AND 0.55 PER CENT MOLYBDENUM, POURED AT 2800°F., TENSILE STRENGTH, 53,700 LB. PER SQ. IN.; BHN, 228; IMPACT, 50 FT.-LB. TOP—UNETCHED, X50. CENTER—ETCHED WITH 2 PER CENT NITAL, X50. BOTTOM—ETCHED WITH 2 PER CENT NITAL, X1000.

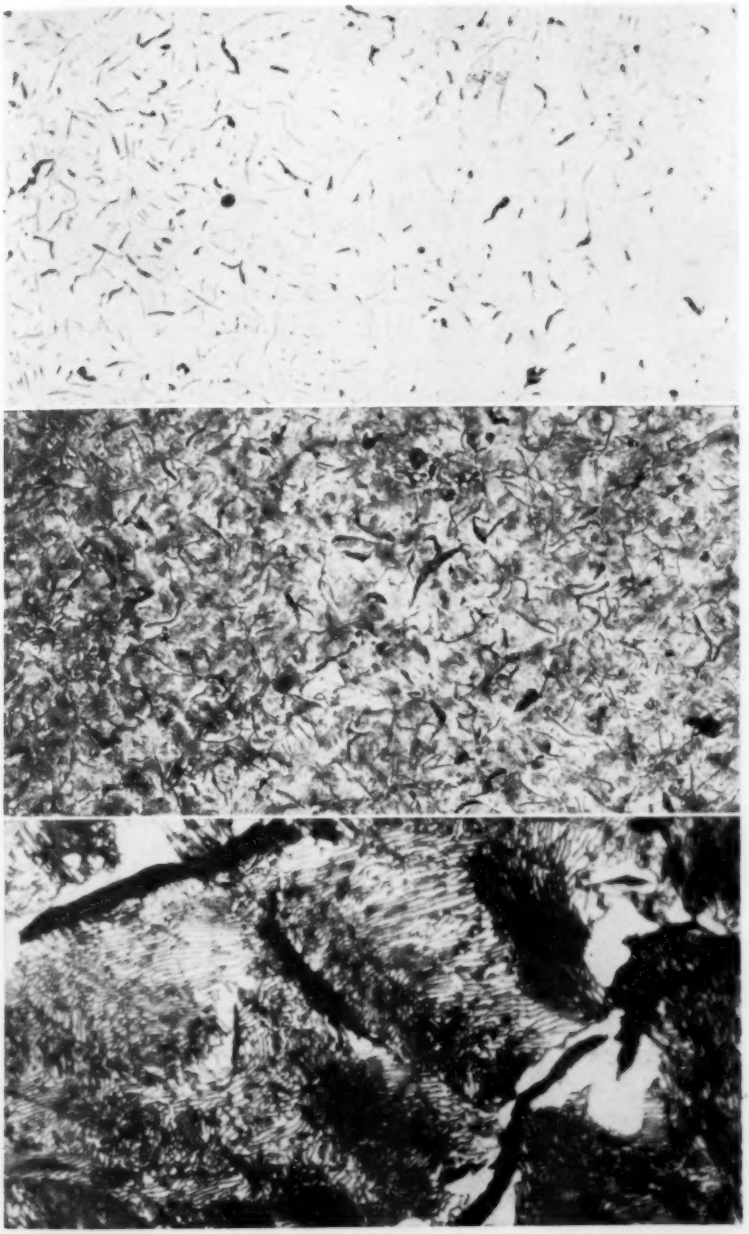


FIG. 2—SAME IRON AS FIG. 1 BUT POURED AT 2650°F.; TENSILE STRENGTH, 50,600 LB. PER SQ. IN.; BHN, 228; IMPACT, 49 FT.-LB. TOP—UNETCHED, X50. CENTER—ETCHED WITH 2 PER CENT NITAL, X50. BOTTOM—ETCHED WITH 2 PER CENT NITAL, X1000.

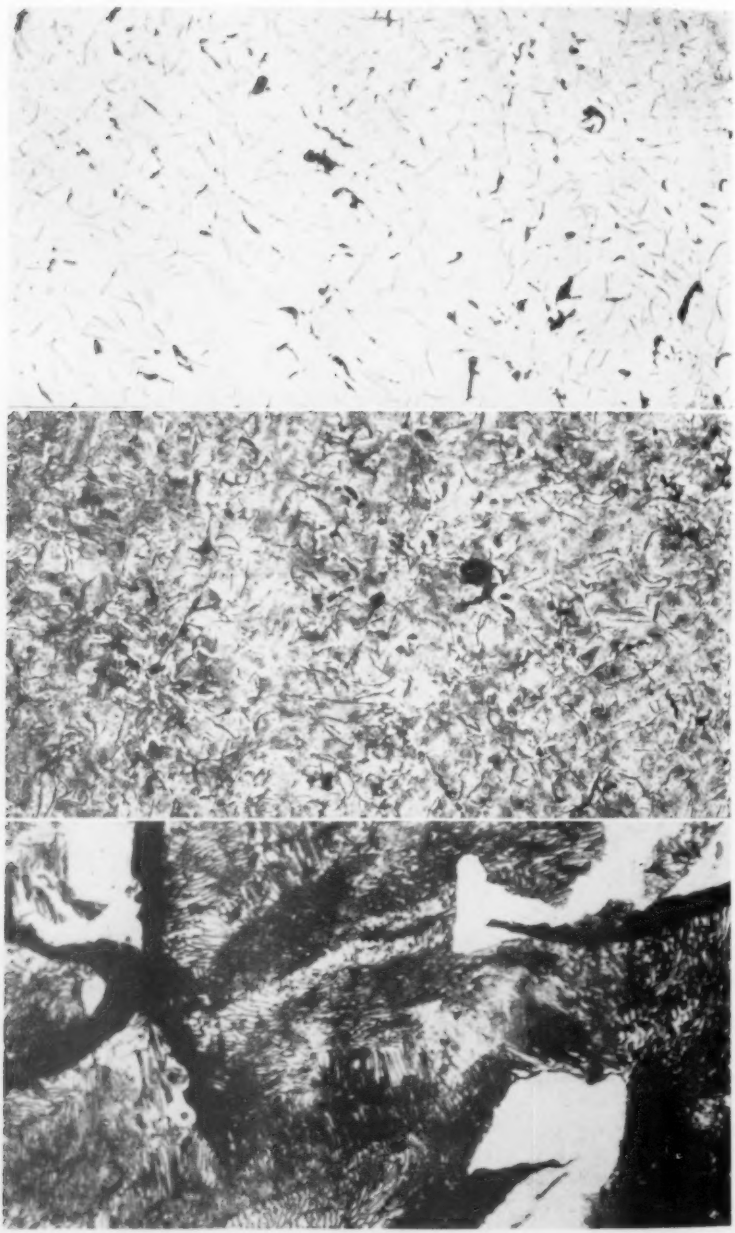


FIG. 3—SAME IRON AS FIG. 1 BUT POURED AT 2500°F.; TENSILE STRENGTH, 45,800 LB. PER SQ. IN.; BHN, 223; IMPACT, 53 FT.-LB. TOP—UNETCHED, X50. CENTER—ETCHED WITH 2 PER CENT NITAL, X50. BOTTOM—ETCHED WITH 2 PER CENT NITAL, X1000.

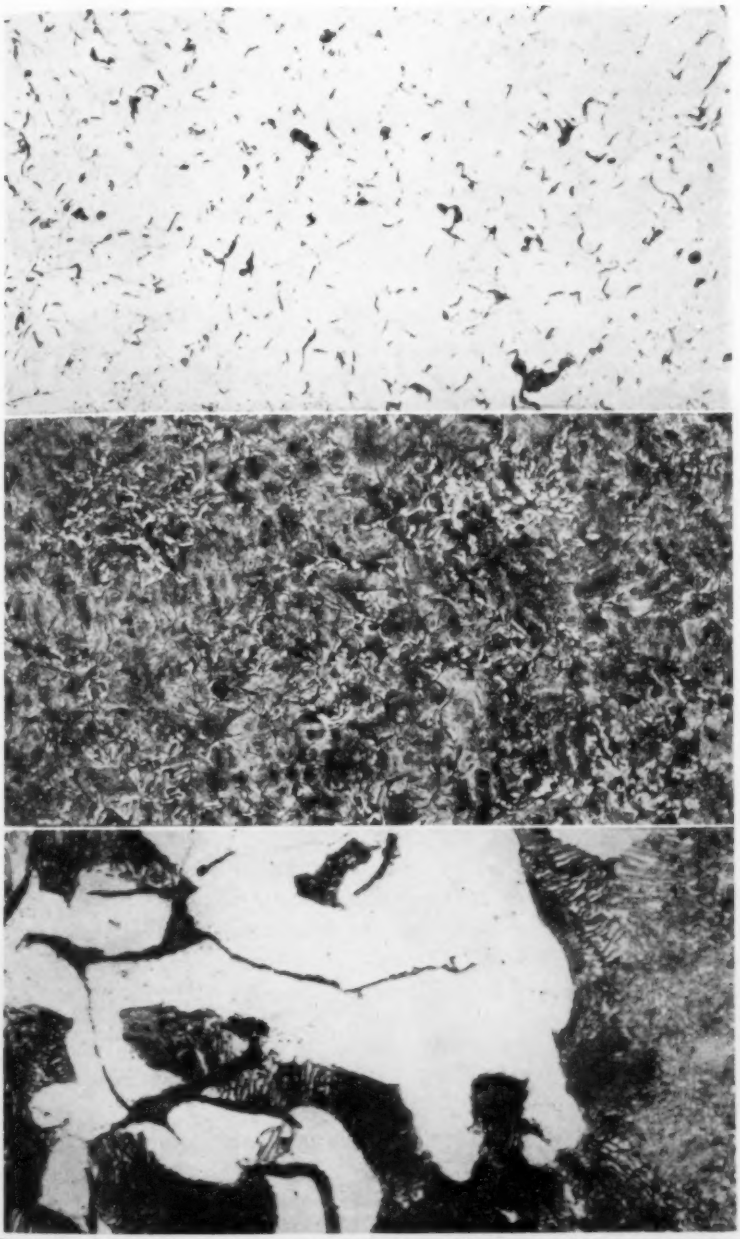


FIG. 4—SAME IRON AS FIG. 1 POURED AT 2350°F. TENSILE STRENGTH, 42,700 LB. PER SQ. IN.; BHN, 217; IMPACT 31 FT.-LB. TOP—UNETCHED, X50. CENTER—ETCHED WITH 2 PER CENT NITAL, X50. BOTTOM—ETCHED WITH 2 PER CENT NITAL, X1000.

Table 3

Heat No.	*1811*	*1810*	*1853*	*2277*	*2279*	*2278*	*1867*	*2280*
Chemical Analyses, per cent								
T.C.	3.49	3.48	3.14	3.00	3.10	3.00	2.92	2.50
Si	2.31	2.14	2.54	2.00	2.20	2.00	1.76	2.75
Mn	0.80	0.86	0.86	0.75	0.75	0.75	0.74	0.80
S			0.083	0.10	0.10	0.10		0.10
P			0.159	0.10	0.10	0.10		0.10
Mo			0.51	0.55	0.50	0.55	0.99	1.00
Cr.					0.50			
Cu						1.00		
Ni								1.00
Pouring Temp., °F.			*Tensile Strength, lb. per sq. in.*					
2800	24,400	30,800	42,500	53,700	53,700	55,400	60,000	66,000
2650	24,900	31,300	41,400	50,600	52,400	55,000	61,200	78,800
2500	24,800	31,500	39,600	45,800	51,500	51,900	58,500	64,700
2350	26,000	30,200	40,000	42,700	42,200	48,400	49,300	48,800
Pouring Temp., °F.				*B. H. N.*				
2800	156	187	212	228	235	241	269	302
2650	160	192	207	228	235	241	277	289
2500	156	187	207	223	235	241	286	285
2350	156	187	207	217	228	248	302	285
Pouring Temp., °F.				*Impact, ft. lb.*				
2800	32	30	35	50	40	51	40	49
2650	33	32	41	49	41	31	35	62
2500	31	34	31	53	55	51	34	51
2350	29	35	30	31	36	45	30	33

NOTE: Chemical analyses given for heats 2277, 2278, 2279, 2280 are "aimed" analyses; the others are actual.

cooled irons and increasing quantities of ferrite definitely decreases the strength of the iron.

16. The impact strengths of the lower tensile strength irons show very little change as the pouring temperature decreases. In iron 1853, the optimum temperature to obtain the maximum impact strength is apparently 2650°F., while in irons 2277, 2279 and 2278 the optimum temperature seems to be 2500°F.

17. Iron 2280, the highly alloyed iron, presents an interesting problem. When poured from 2800°F., the cooling rate seems to be sufficiently slow to permit the formation of a large amount of pearlite. The tensile and impact strengths are lower when compared with the same iron poured at a temperature of 2650°F. The test bar poured at the latter temperature was almost completely acicular in the matrix microstructure. This structure is usually associated with high tensile strength and high impact strength.

18. When the pouring temperature is reduced to 2500°F., the graphite formation tends more toward the whorl type of pattern and the quantity of ferrite is slightly increased. Both the tensile strength and the impact resistance are reduced.

19. With a pouring temperature of 2350°F., larger areas of ferrite become apparent in the matrix of even this highly alloyed iron, and the whorl type graphite distribution becomes more prominent. These changes are probably responsible for the further reduction of tensile and impact strengths.

20. Bars poured at 2500 and 2350°F. contained numerous gas holes near the surface. These holes were of the order of magnitude of spheres $\frac{1}{8}$-$\frac{3}{16}$-in. diameter. Some of the holes contained small globules of metal, a manifestation of liquation.

21. The increased precipitation of ferrite in test bars poured at lower temperatures prevailed when heats were melted down with a charge consisting of all steel and when the charge consisted of 50-75 per cent chateguay pig. The use of late additions of ferrosilicon in heats 1810, 1811, 1853, 1867 did not alter this general trend.

22. Following the changes in microstructures with the change in pouring temperature of the gray irons herein considered, it is found that, as the pouring temperature is reduced, the graphite tends to become more dendritic in nature, approaching the eutectiform pattern in some cases. This change in graphite distribution indicates that the lower pouring temperatures are conducive to undercooling.[10] Considering the effect of pouring temperature on the cooling rate, the theory of undercooling is substantiated.

23. The molds used, were all of baked sand and contained the same volume and weight of sand around each test bar which should provide a uniform insulation for the set of bars. If a high pouring temperature was used, the iron poured into the mold should have possessed a higher heat content than when the same iron was cast at a lower temperature. Since all the heat must be dissipated through the baked sand insulation, and since this insulation, is uniform from bar to bar, the time required to remove the total heat content must be longer in the case of the bar poured at a higher temperature and the cooling rate must be slower. It is well known that a higher rate of cooling tends to promote undercooling.

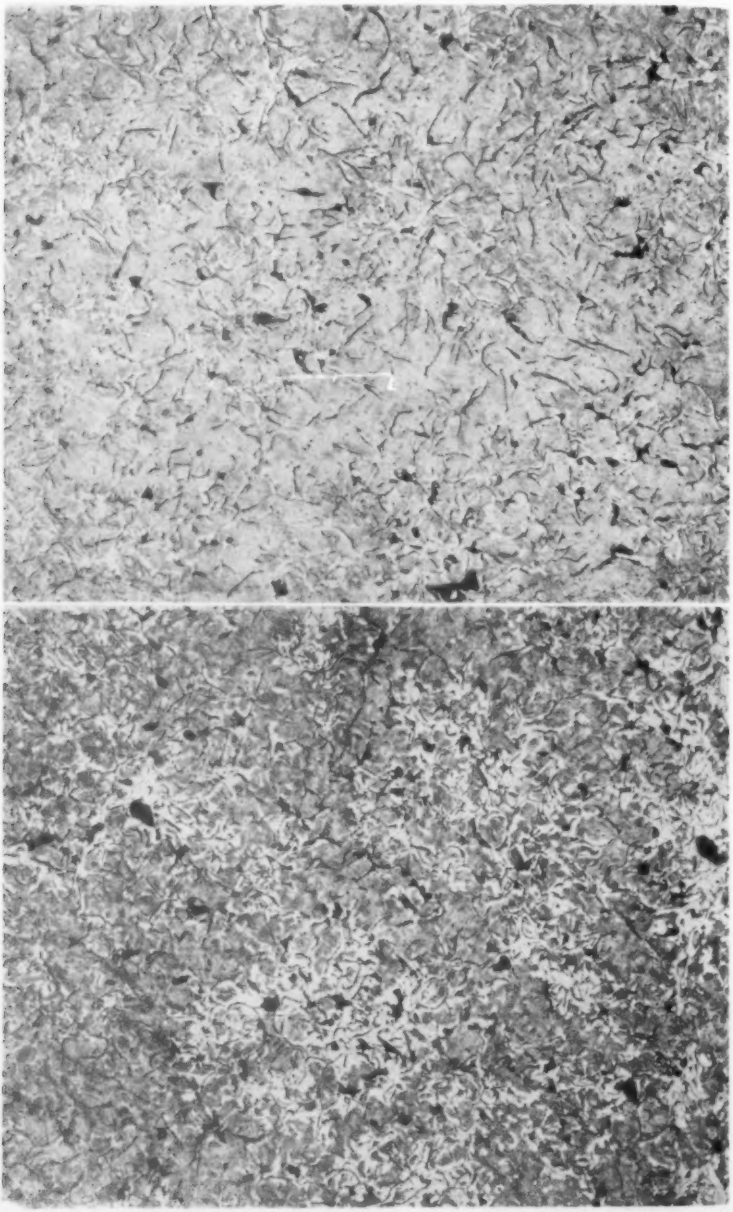


FIG. 5—HEAT 1811 CONTAINING 3.49 PER CENT TOTAL CARBON, 2.31 PER CENT SILICON, 0.80 PER CENT MANGANESE. TOP—POURED AT 2800°F.; TENSILE STRENGTH, 24,400 LB. PER SQ. IN.; BHN—156; IMPACT, 32 FT.-LB. ETCHED IN 2 PER CENT NITAL, X50. BOTTOM—POURED AT 2650°F.; TENSILE STRENGTH, 24,900 LB. PER SQ. IN.; BHN, 160; IMPACT, 33 FT.-LB. ETCHED IN 2 PER CENT NITAL, X50.

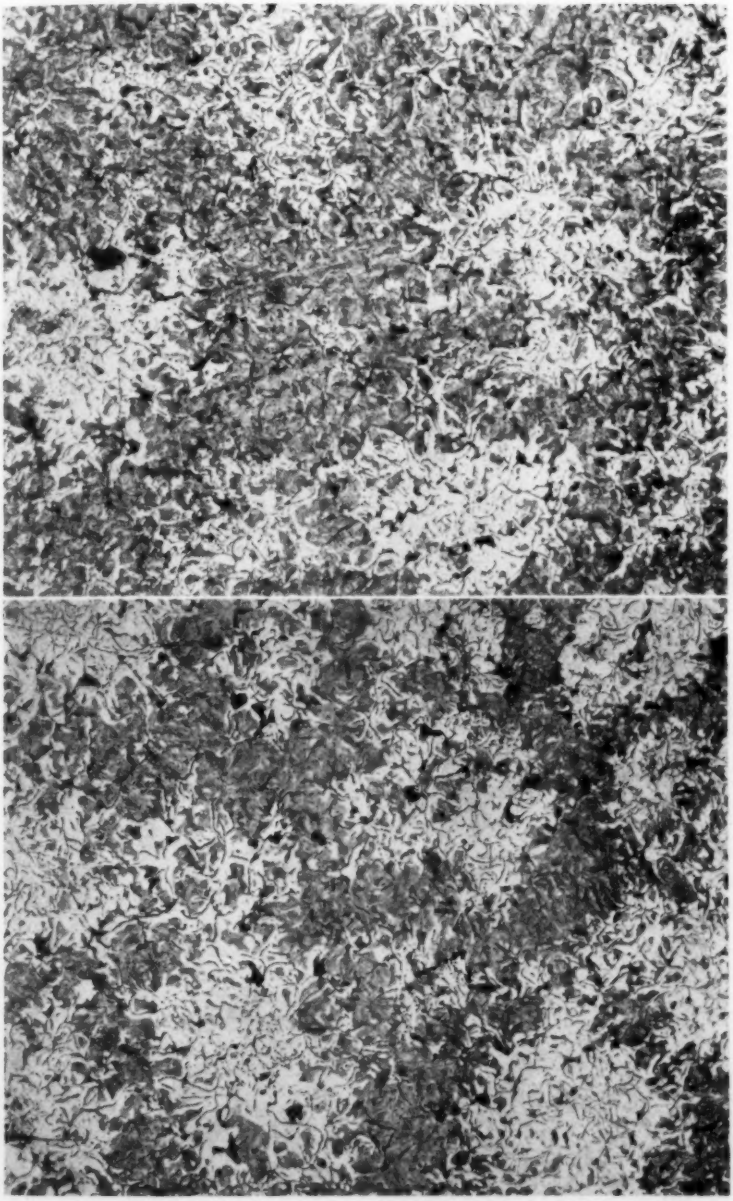


FIG. 6—SAME IRON AS FIG. 5. TOP—POURED AT 2500°F.; TENSILE STRENGTH, 24,800 LB. PER SQ. IN.; BHN, 156; IMPACT, 31 FT.-LB. ETCHED IN 2 PER CENT NITAL, X50. BOTTOM—POURED AT 2350°F.; TENSILE STRENGTH, 26,000 LB. PER SQ. IN.; BHN, 156; IMPACT, 29 FT.-LB. ETCHED IN 2 PER CENT NITAL, X50.

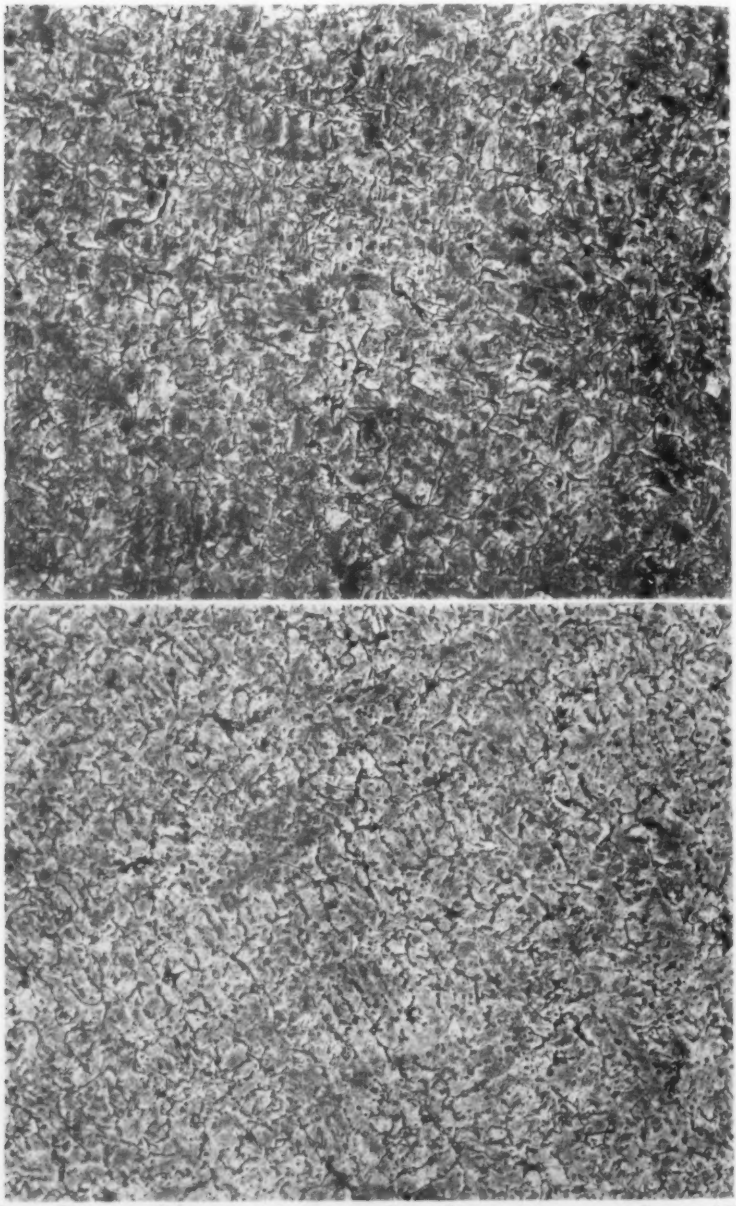


FIG. 7—HEAT 1867 CONTAINING 2.92 PER CENT TOTAL CARBON, 1.76 PER CENT SILICON, 0.74 PER CENT MANGANESE, AND 0.99 PER CENT MOLYBDENUM. TOP—POURED AT 2800°F.; TENSILE STRENGTH, 60,000 LB. PER SQ. IN.; BHN, 269; IMPACT, 40 FT.-LB. ETCHED IN 2 PER CENT NITAL, X50. BOTTOM—POURED AT 2650°F.; TENSILE STRENGTH, 61,200 LB. PER SQ. IN.; BHN, 277, IMPACT, 35 FT.-LB. ETCHED IN 2 PER CENT NITAL, X50.

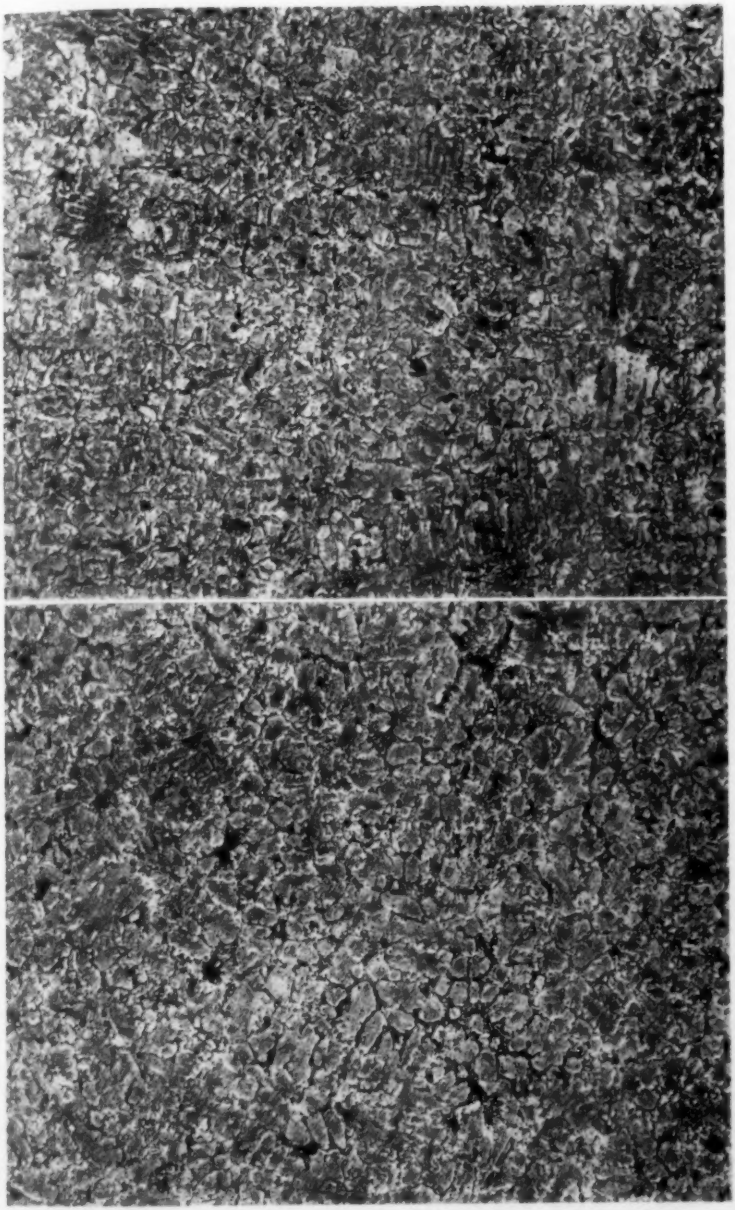


FIG. 8—SAME IRON AS FIG. 7. TOP—POURED AT 2500°F.; TENSILE STRENGTH, 58,500 LB. PER SQ. IN.; BHN, 286; IMPACT, 34 FT.-LB. ETCHED IN 2 PER CENT NITAL, X50. BOTTOM—POURED AT 2350°F. TENSILE STRENGTH, 49,300 LB. PER. SQ. IN.; BHN, 302; IMPACT, 30 FT.-LB. ETCHED IN 2 PER CENT NITAL, X50.

24. It is recognized that there are probably other factors influencing undercooling, such as composition and nucleation. DiGiulio and White[1] have shown a relation between superheating and graphite arrangement and based their conclusions upon nucleation phenomenon. The works of Bancroft and Dierker[6] and Parke, Herzig and Crosby[7] have revealed the effect of undercooling (or rate of solidification) on graphite arrangement and they showed an increased quantity of ferrite associated with the undercooled graphite pattern.

25. Consideration of the association of ferrite with graphite distributions, indicative of undercooling prior to solidification, or faster rates of cooling, in the light of recent investigations concerned with the transformations taking place in gray iron as it cools from the casting temperature to room temperature, suggests that the phenomenon is unexpected and may be due to some factor which heretofor has not received much attention. Boyles[8] has shown that no ferrite exists in irons of usual compositions above the pearlite interval. Under equilibrium conditions, ferrite is formed within the pearlite interval from the decomposition of austenite. In his work, it was revealed that austenite, ferrite and graphite coexist within this temperature interval and that the percentage of austenite decreases to zero at the lower limit, causing the matrix to become completely ferritic, under equilibrium conditions, by the time this lower limit has been reached.

26. For most irons, however, equilibrium is never obtained, and the high strength pearlitic irons owe their physical properties to a certain amount of austenite stability which prevents precipitation of ferrite as the iron cools in the mold from the casting temperature to room temperature. Boyles has also indicated that with certain degrees of undercooling, ferrite precipitates at temperatures well below the pearlite interval as the first product of austenite decomposition. For the iron studied, a maximum rate of ferrite precipitation was found at about 1350°F. Murphy, Wood and D'Amico[9], in dealing with austenite transformation in gray iron, showed ferrite precipitation adjacent to graphite flakes in samples isothermally transformed at temperatures as low as 1100°F.

27. While it is recognized that ferrite is the result of austenite decomposition under equilibrium (or near equilibrium) con-

ditions, it is not readily understood why a relatively slowly cooled casting containing random graphite distributions should have a matrix of pearlite with no ferrite present, while a more rapidly cooled iron, subjected to considerable undercooling, should contain ferrite in the matrix with graphite distributions of the whorl, dendritic and eutectiform types.

28. Whenever ferrite occurs in cast iron it lies adjacent to graphite flakes. Since it has been shown that this ferrite results from the decomposition of austenite, a rapid rate of diffusion of carbon in alpha iron is implied. Microstructures of samples quenched at various temperatures during furnace cooling indicate that, at the start of transformation, a small amount of ferrite is first formed adjacent to a graphite flake and, as cooling progresses (or during a continued sojourn at the temperature in the range where ferrite is a product of austenite decomposition) this ferrite envelope increases in width. The consensus is that austenite gives up its carbon to the nearest graphite flake, the original austenite grain decreasing in size as the ferrite grains become larger. However, in order for the austenite to give up its carbon to this flake, the carbon must diffuse through the already formed ferrite layer. When the temperature is further reduced, the remaining austenite grain transforms to pearlite.

29. In comparing the rates of ferrite formation at various temperatures in a normal iron with those of a "modified" iron (eutectiform graphite distribution), Boyles[8] found that the latter iron exhibited the faster rate at each temperature, and attributed this increased rate to a larger graphite-austenite interface. However, a study of his photomicrographs revealed that, in several cases, the ferrite layers separating pearlite from graphite flakes in the modified iron were wider than the ferrite layers found in the normal iron, indicating that during isothermal transformation at the same temperature the distance between the austenite-ferrite interface and the ferrite-graphite interface was greater in the modified iron than in the normal iron. Therefore, it seems probable that the modified iron may have contained some element which would tend to increase the rate of diffusion of carbon through alpha iron.

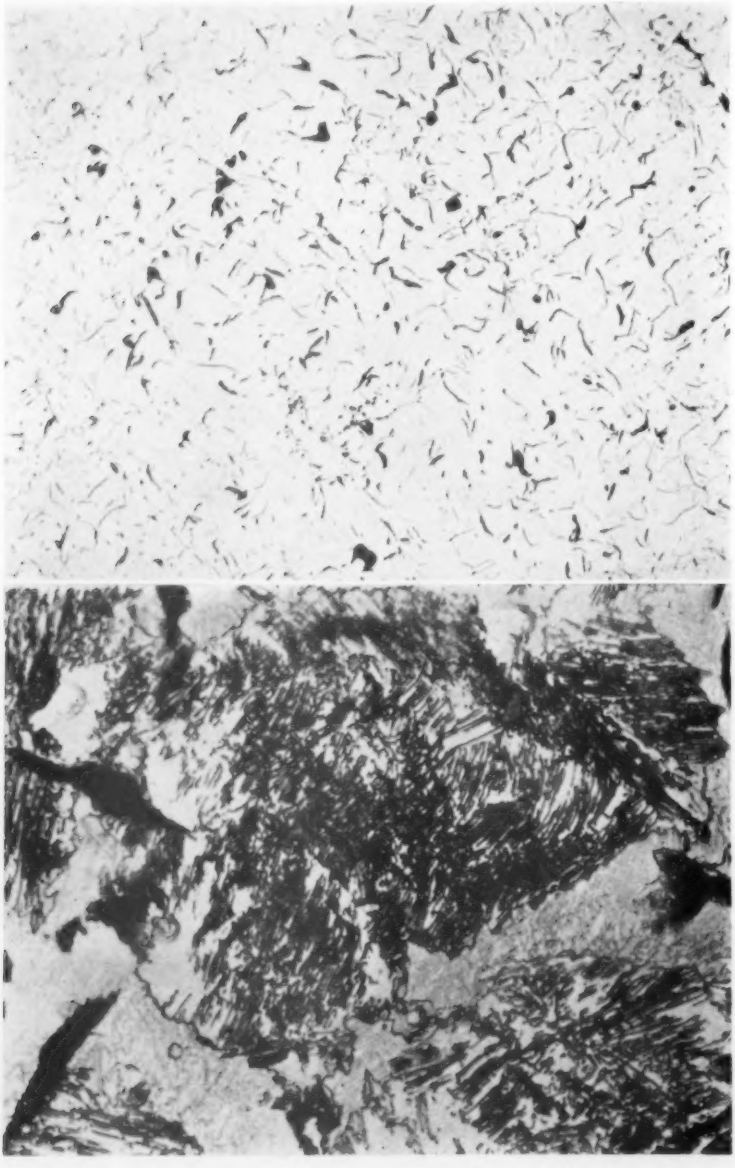


FIG. 9—IRON 2280 CONTAINING 2.50 PER CENT TOTAL CARBON, 2.75 PER CENT SILICON, 0.80 PER CENT MANGANESE; 0.10 PER CENT PHOSPHORUS, 0.10 PER CENT SULPHUR, 1.00 PER CENT MOLYBDENUM, AND 1.00 PER CENT NICKEL. POURING TEMPERATURE, 2800°F.; BHN, 302; TENSILE STRENGTH, 66,000 LB. PER SQ. IN. TOP—UNETCHED, X100. BOTTOM—ETCHED WITH 2 PER CENT NITAL, X1000.

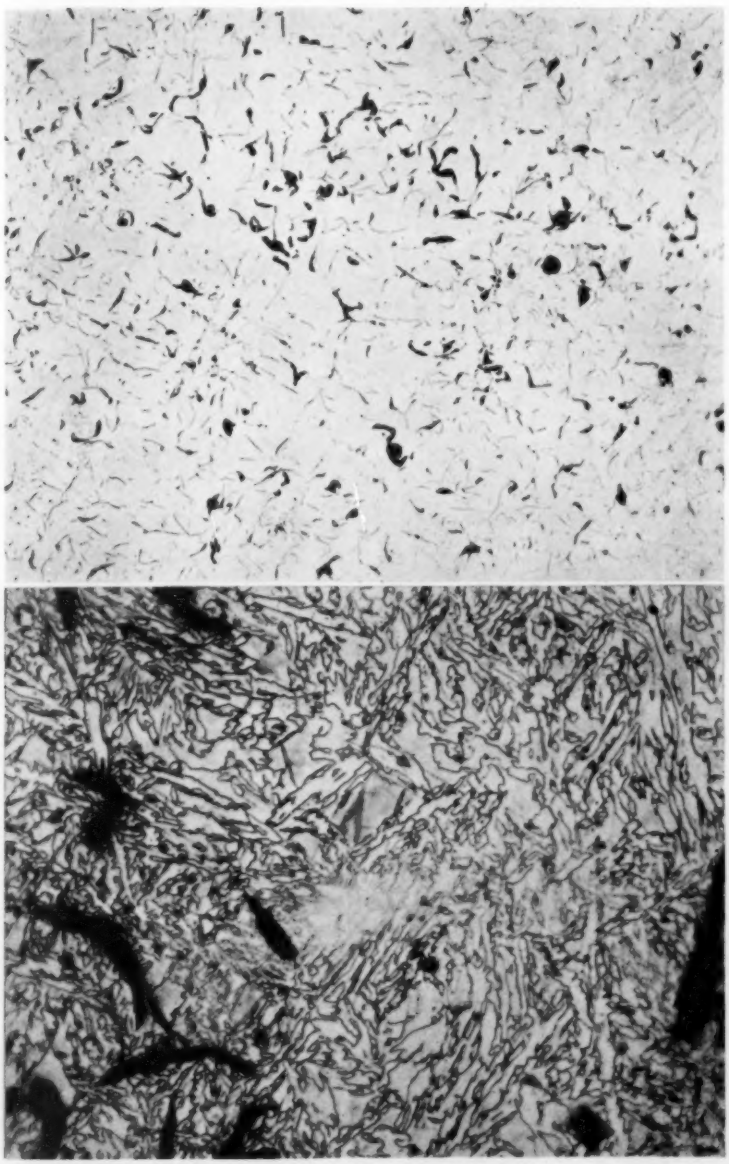


FIG. 10—SAME IRON AS FIG. 9 BUT POURED AT 2650°F.; BHN, 285; TENSILE STRENGTH, 78,800 LB. PER SQ. IN. TOP—UNETCHED, X100. BOTTOM—ETCHED WITH 2 PER CENT NITAL, X1000.

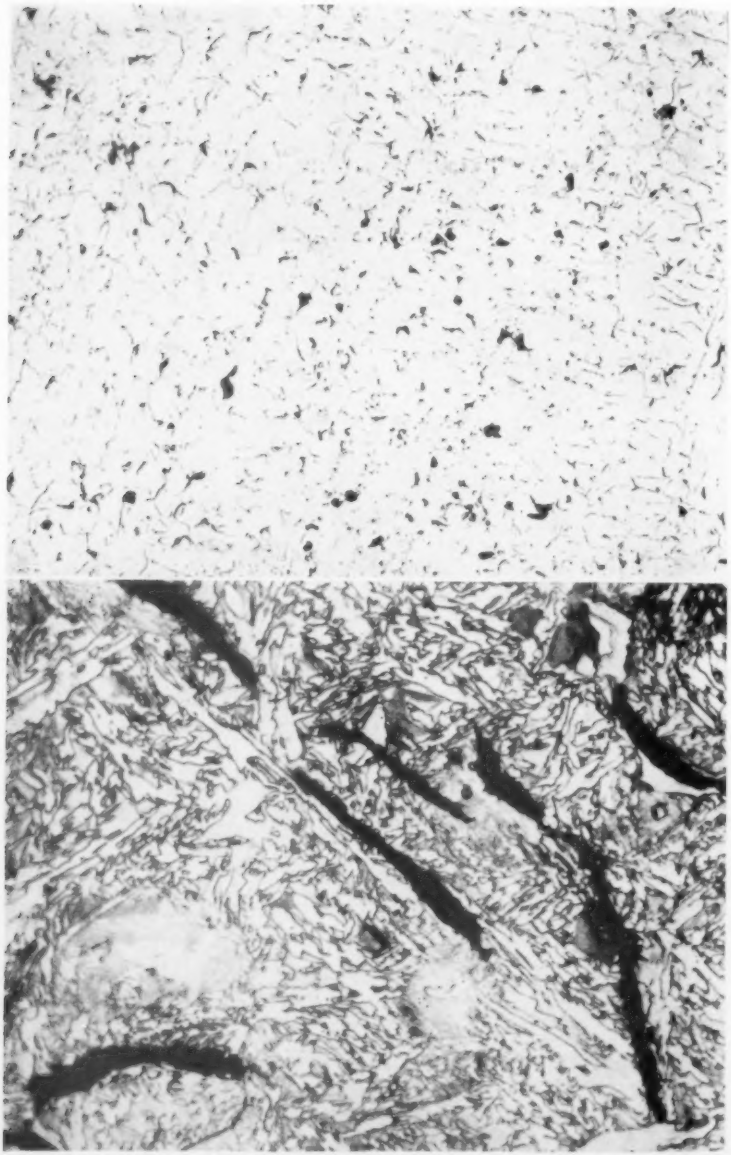


FIG. 11—SAME IRON AS FIG. 9 BUT POURED AT 2500°F.; BHN, 285; TENSILE STRENGTH, 64,700 LB. PER SQ. IN. TOP—UNETCHED, X100. BOTTOM—ETCHED WITH 2 PER CENT NITAL, X1000.

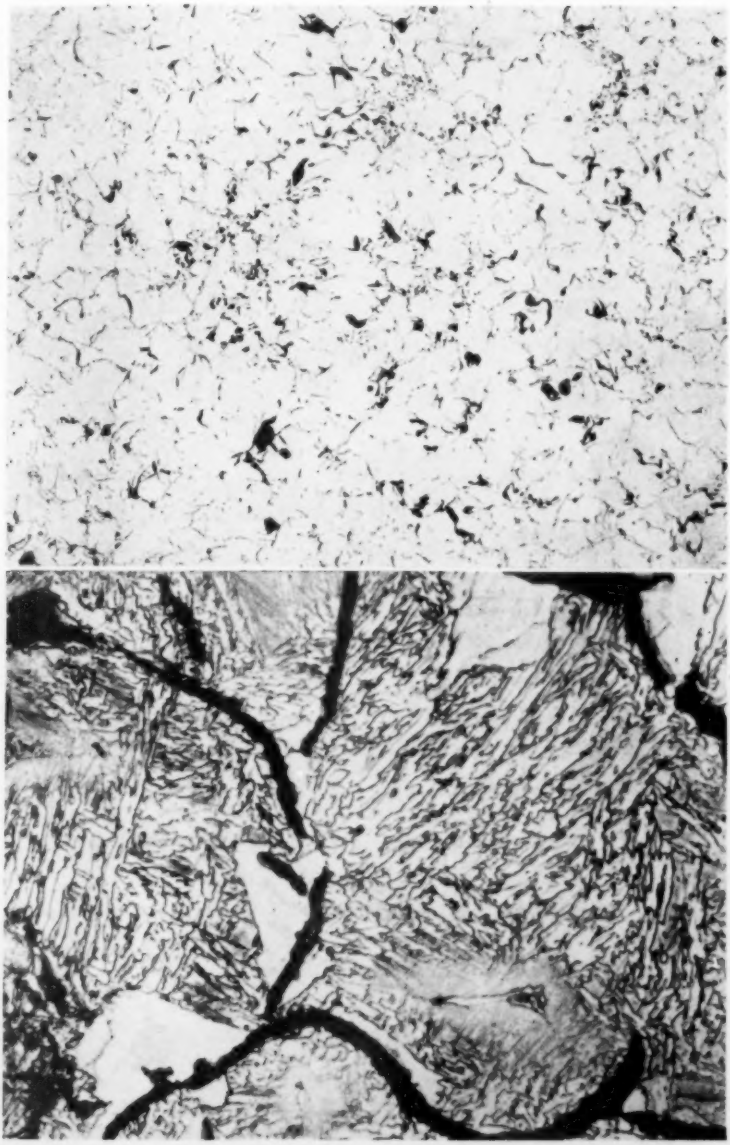


FIG. 12—SAME IRON AS FIG. 9 BUT POURED AT 2350°F.; BHN, 285; TENSILE STRENGTH, 48,800 LB. PER SQ. IN. TOP—UNETCHED, X100. BOTTOM—ETCHED WITH 2 PER CENT NITAL, X1000.

A POSSIBLE EXPLANATION FOR THE PRESENCE OF FERRITE IN IRONS POURED AT LOWER TEMPERATURES

30. The eutectiform and whorl types of graphite distribution have been found by Schneidewind and D'Amico[10] to result from undercooling. Their work related the type of graphite distribution to the temperature of solidification and experiments in isothermal solidification of cast iron at temperatures below the equilibrium eutectic temperature were conducted to illustrate this relationship. The association of ferrite with these types of graphite distribution in the test bars poured at lower temperatures, even though these bars were subjected to a faster rate of cooling from the casting temperature to room temperature, suggests that ferrite is related to undercooling prior to solidification.

31. The formation of a layer of ferrite between graphite and pearlite is similar in nature to the microstructure of a chilled gray iron. This latter structure often consists of carbide and pearlite, the latter containing coarse carbide plates and sometimes spheroidized carbides, and separated from the cementite by ferrite (See Fig. 14). This is an unusual and unexpected structure when the rate of cooling of the chilled section is compared with the cooling of a heavier section of the same composition which contains graphite in a matrix of pearlite.

32. Bancroft and Dierker[6], and Parke, Crosby and Herzig[7] concluded that for every cast iron there existed a certain critical cooling rate which was conducive to the formation of the eutectiform type graphite distribution associated with ferrite. The eutectiform and whorl types of graphite distribution have been explained by the undercooling theory referred to above. No explanation or theory has been advanced regarding the formation of ferrite in a zone subjected to rapid cooling while a heavier, more slowly cooled section contained a pearlitic matrix, or the presence of ferrite in test bar sections poured at temperatures from 2500-2300°F. Any such explanation must consider the rapid diffusion of carbon through alpha iron.

33. Abnormality in steels has been recognized since the development of the McQuaid-Ehn grain size test. The term applies to the microstructure of the hypereutectoid case of box-cooled, carburized specimens of certain steels. Bain[11] has described the abnormal structure as follows: "thick carbide envelopes with

coarse lamellar pearlite, if any, separated from the network, which may be discontinuous, by a wide band of ferrite sometimes containing large spheroids of carbide."

34. Keeping the conditions of carburizing constant and holding the cooling rates to the same speed, various steels behave differently with regard to the amount of ferrite formed between carbide and pearlite in the hypereutectoid zone and the differences have been attributed to variations in the steel melting and "killing" processes involved in the production of the heats. "Normal" steels consist of pearlite and the carbide network without any free ferrite, while the "abnormal" steels consist of pearlite, the ferrite band, and the carbide network. It has been shown that the ferrite layer forms by the decomposition of austenite and involves the diffusion of carbon from an austenite grain, through the first formed ferrite layer to the carbide network.

35. The abnormal structure contains the same microconstituents which are found in chilled iron and the relative positions of these microconstituents are the same in each case (See Figs. 13 and 14). The presence of ferrite between pearlite and graphite in undercooled structures of gray cast iron (and, of course, the struc-

FIG. 13—STRUCTURE OF ABNORMAL STEEL, ETCHED WITH 2 PER CENT NITAL, X1000.

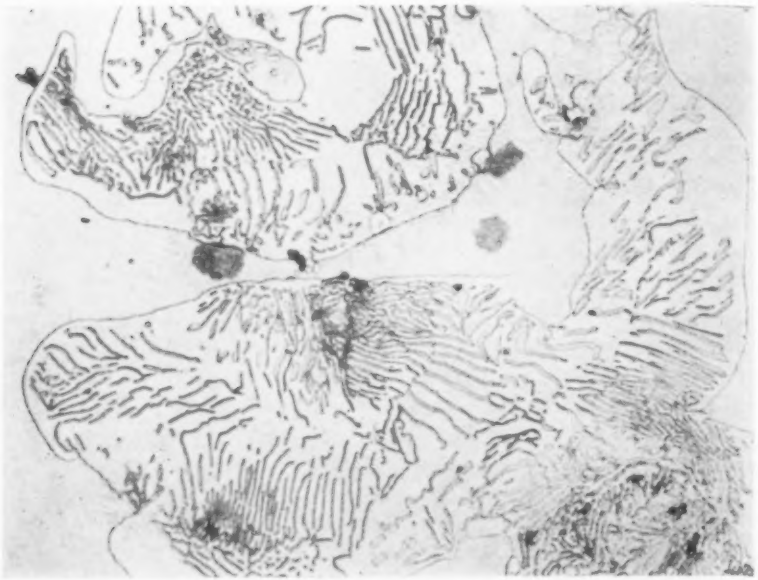


FIG. 14—STRUCTURE OF CHILLED IRON, ETCHED WITH 2 PER CENT NITAL, X1000.

ture found in unalloyed heavy sections of gray iron which cooled so slowly that reactions in the solid state took place under near-equilibrium conditions) is similar to the presence of ferrite in the other two structures, except that graphite is found in place of cementite. It is apparent that the three structures are related in some manner and that this relationship depends upon rapid diffusion of carbon in ferrite, which may be attributed to the same cause in each case.

36. To produce these structures, it is essential that the austenite possess a low stability or a transformation rate sufficiently fast to prevent undercooling to temperatures of slow carbide diffusion, and a sufficiently rapid diffusivity of carbon in alpha iron at temperatures in the range 1250-1400°F. to effect carbon migration to the carbide or graphite more rapidly than pearlite can be formed.

37. Since, during constant cooling, the metal can remain in the temperature range conducive to ferrite formation a limited length of time, and since the austenite grains may be fairly large, some of the austenite eventually transforms to pearlite and the final structure consists of graphite, or carbide, plus ferrite and pearlite.

38. The phenomenon of abnormality in steels has been studied by many investigators. It has generally been associated with dissolved oxygen [11, 12, 13, 14, 15] in steels and sometimes with nitrogen[16-17]. In either case, it has been conceded that the presence of some dissolved gas is instrumental in the formation of this structure. Since cast iron contains relatively large quantities of carbon and silicon, it is difficult to believe that oxygen could be the element responsible for the rapid diffusion of carbon in ferrite. Boyles[18, 19] has investigated the effect of oxygen on the graphitization of gray cast iron and showed that dissolved oxygen was not responsible for the precipitation of ferrite. However, some experimental heats melted in an atmosphere of nitrogen exhibited structures containing ferrite adjacent to the graphite flakes of furnace cooled samples.

39. There is evidence that nitrogen does increase the diffusion of carbon in alpha iron. Beside the experiment of Boyles, investigators[20] dealing with arc welding in an atmosphere of nitrogen have noted that an abnormal structure resulted from carburizing the weld metal. Others[16, 17] have also noted an influence of nitrogen upon the formation of the abnormal structure.

40. It is known that nitrogen is more soluble in liquid iron than in solid iron and that a sharp increase in solubility takes place at the melting point. Undercooling, therefore, may produce a supersaturated solution of nitrogen in iron which, at temperatures where alpha iron is the equilibrium microconstituent, increases the rate of carbon diffusion and promotes the formation of ferrite in chilled iron or in iron undercooled to the extent that the eutectiform graphite pattern is formed upon solidification. Upon cooling more slowly, the gas liberated during solidification is allowed to escape and the slow diffusion of carbon in alpha iron without the presence of nitrogen prevents, or reduces to a minimum, the quantity of ferrite formed.

41. The presence of the gas holes at the surfaces of the test bars cast at lower temperatures suggests the presence of an excess of gas during solidification, whereas the bars cast at higher temperatures were sound all the way through.

42. As further support for this theory, the following correlated data are presented from the literature. In the work of DiGiulio and White, a table concerned with "Gas Content of

Superheated Gray Cast Iron as Determined by the Vacuum Fusion Method'' showed that an iron containing 3.21 per cent carbon and 2.24 per cent silicon, superheated to a temperature of 2855°F., contained 0.0073 per cent nitrogen after cooling to room temperature. In U. S. patent No. 2,229,140, dealing with abnormal steel, it was found that a steel containing 0.007-0.012 per cent nitrogen was abnormal when carburized according to McQuaid-Ehn grain size specifications.

CONCLUSIONS

43. Data have been presented which lead to the following conclusions:

1. As the pouring temperature is lowered, the graphite pattern becomes finer and tends toward the dendritic, whorl and eutectiform types of distribution.

2. As the pouring temperature is lowered, the quantity of ferrite precipitated during the gamma to alpha phase change increases.

3. For irons of relatively low tensile strengths, the increase in ferrite produced by a reduction in the pouring temperature apparently has little effect upon the tensile or impact properties of the iron.

4. For higher strength irons in the class often referred to as high test irons, the effect of lowering the pouring temperature is pronounced and the strengths are reduced considerably by reducing the pouring temperature to below 2650°F.

5. The presence of ferrite in even highly alloyed irons has been found to be influenced by reduced pouring temperature.

6. Late additions of silicon were found to have little effect upon the precipitation of ferrite in bars poured at lower temperatures.

7. The nature of the charge—whether all steel or only 25-50 per cent steel (rest pig iron) — apparently had little effect upon ferrite precipitation in castings poured at lower temperatures.

44. Based upon these conclusions, it is recommended that gray cast iron be poured at temperatures above 2600°F. when the presence of ferrite in a casting is undesirable. Photomicrographs have been presented to show the trend in changes of microstructure effected by a reduction of pouring temperature.

45. Some relation between abnormality in steel, the microstructure of chilled cast iron and the formation of ferrite in gray cast iron has been suggested, since each microstructure involves the rapid diffusion of carbon in alpha iron.

Bibliography

1. DiGiulio, A. and White, A. E., *"Factors Affecting the Structure and Properties of Gray Cast Iron,"* TRANSACTIONS, American Foundrymen's Association, vol. 43, p. 531 (1935).

2. Bolton, J. W., *"Graphitization and Inclusions in Gray Iron,"* TRANSACTIONS, American Foundrymen's Association, vol. 45, p. 467 (1937).

3. Roth, E. L., *"Sixty Thousand Pound Per Square Inch Cupola Iron,"* TRANSACTIONS, American Foundrymen's Association, vol. 47, pp. 873 and 877 (1939).

4. Bardenheuer and Zeyen, ALLOYS OF IRON AND CARBON, VOL. II, p. 324.

5. Ziegler, N. A., and Northrup, H. W., *"Effect of Superheating on Castability and Physical Properties of Cast Irons of Different Carbon Contents,"* TRANSACTIONS, American Foundrymen's Association, vol. 47, p. 620 (1939).

6. Bancroft, R. H., and Dierker, A. H., *"Ferrite—Its Occurrence and Control in Gray Cast Iron,"* TRANSACTIONS, American Foundrymen's Association, vol. 45, p. 449 (1937).

7. Parke, R. M., Crosby, V. A., and Herzig, A. J., *"Notes on the Graphitization of Gray Cast Iron,"* METALS & ALLOYS, vol. 9, p. 9, January 1938.

8. Boyles, Alfred, *"The Pearlite Interval in Gray Cast Iron,"* TRANSACTIONS, American Foundrymen's Association, vol. 48, p. 531 (1941).

9. Murphy, D. W., Wood, W. P., and D'Amico, C. D., *"Austenite Transformation in Gray Iron,"* TRANSACTIONS, American Foundrymen's Association, vol. 46, p. 563 (1938).

10. Schneidewind, R., and D'Amico, C. D., *"The Influence of Undercooling on the Graphite Pattern of Gray Cast Iron,"* TRANSACTIONS, American Foundrymen's Association, vol. 47, p. 831 (1939).

11. Bain, E. C., and Davenport, E. S., *"General Relations Between Grain-Size and Hardenability and the Normality of Steels,"* TRANSACTIONS, A.S.M., vol. 22, p. 879 (1934).

12. Derge, G., Kommel, A. R., and Mehl, R. F., *"Some Factors Influencing Austenitic Grain Size in High Purity Steels,"* TRANSACTIONS, A.S.M., vol. 26, p. 153 (1938).

13. Grossmann, M. A., *"On Oxygen Dissolved in Steel, and Its Influence on the Structure,"* TRANSACTIONS, A.S.M., vol. 16, p. 1, (1929).

14. Grossmann, M. A., *"On Grain Size and Grain Growth,"* TRANSACTIONS, A.S.M., vol. 21, p. 1079 (1933).

15. Bain, E. C., *"Factors Affecting the Inherent Hardenability of Steel,"* TRANSACTIONS, A.S.S.T., vol. 20, p. 385 (1932).

16. U. S. Patent No. 2,229,140—"Abnormal Steel," Earle C. Smith and George T. Motok.

17. Reeder and Brophy, *"Nitrogen in Carburized Steels,"* Chem. & Met. Eng., 25:867 (1921).

18. Boyles, Alfred, *"The Formation of Graphite in Gray Iron,"* TRANSACTIONS, American Foundrymen's Association, vol. 46, p. 297 (1938).

19. Boyles, Alfred, *"Influence of Atmosphere and Pressure on Structures of Iron-Carbon-Silicon Alloys,"* TRANSACTIONS, A.I.M.E., vol. 135, p. 376 (1939).

20. Weber, Ludwig J., *"Studies on Electric Welding,"* TRANSACTIONS, A.S.S.T., vol. 11, p. 425 (1927).

DISCUSSION

Presiding: J. W. BOLTON, Lunkenheimer Co., Cincinnati, Ohio.

CHAIRMAN BOLTON: I might point out that our own experience differs from that of Messrs. Timmons and Crosby. If you lower the pouring temperature sufficiently with the so-called lower strength irons, you will get into exactly the same set of conditions he gets into with the higher ones. If, on the other hand, you compare them all on the same base level of temperature, as apparently has been done in this investigation, you may find just what the authors have found. In other words, the lower strength iron may stand a lower temperature of pouring, but such iron has a minimum safe range. The authors were within the safe range in their experiments for the lower strength irons.

J. S. VANICK[1]: I would have a hard time trying to remember a very intricate presentation of this kind but I would like to leave the matter of

[1] International Nickel Co., Inc., New York, N. Y.

graphite and ferrite formation to the paper, presented by Mr. Eash, which is shown elsewhere in this volume*.

What strikes me at this time is, Mr. Timmons' experiments did not seem to show any particular advantage for late silicon additions and the reason must have been either that the inoculant was ineffective, or the quantities used were inadequate; or, due to some other factor, they failed to do the job properly, and a more effective inoculating practice would seem to be essential. Inoculation certainly must be worked into a safer practice for the foundryman if differences in pouring temperatures, that is, dropping down to as low as 2550°F., let us say, knocks the strength of a 77,000 lb. per sq. in. iron down to something like 60,000 lb. per sq. in. The foundryman must have a more secure way of controlling that product than to try to hold a temperature that he cannot control after the metal gets into the ladle.

In the matter of using austenite stabilizers to prevent ferrite precipitation, some interesting thought might be spent upon these compositions, with the object of working out balances which will adjust these compositions to possess the proper, or as nearly as possible a proper, concentration to convert from austenite to the acicular structure without the precipitation of ferrite, and in that way again control or maintain, the high strength.

W. A. PENNINGTON[2]: I should like to add one thing, to what Mr. Timmons has said, in connection with his implication that carbon diffuses rather rapidly through alpha iron. It seems to me to be fairly well proven that that is just the case. The contention is substantiated by certain things which happen in decarburization, which I think is quite closely related to graphitization, anyway.

If an attempt is made to decarburize a low carbon steel by heat treatment, it will be found that the decarburization proceeds faster around 1400 to 1450°F. than it does in the austenitic range around 1700 to 1750°F. In both cases, the carbon diffuses through the iron in solid solution, to the surface. The diffusion is through alpha iron at the lower and through gamma iron at the higher temperature. Since the reaction goes faster at the lower temperature, it may be assumed that the diffusion of carbon through alpha iron at 1450°F. is faster than the diffusion through gamma iron at 1750°F. with the concentration gradients of carbon imposed by the above conditions.

There is one other thing that I should like to mention in connection with this paper that may later lend some bearing to the solution of certain problems. In annealing sheet steels in the presence of mill scale in the so-called box anneal, it will be found that there is no appreciable decarburization until some 1275 to 1300°F. has been reached. On increasing the temperature from this point, a range is passed in which the decarburization is quite fast. If the proper composition be selected,

* Eash, J. T., "Effect of Ladle Inoculation on the Solidification of Gray Cast Iron."
[2] Mellon Institute of Industrial Research, Pittsburgh, Pa.

the decarburization at 1600°F. will cease after a short time due to the building up of an equilibrium atmosphere. Of course, there is an equilibrium atmosphere and carbon content of the steel at 1450°F., but it cannot be reached until the carbon in the steel has been reduced to some 0.015 per cent, providing the pressure is in the neighborhood of one atmosphere. A further increase in temperature to above A_3 point for pure iron assures faster and more complete decarburization than is had at 1600°F. The end-point carbon, as in the case of the 1450°F. anneal, is low.

So as the temperature is raised, there is practically no decarburization up to 1275°F., then decarburization through a range up to some 1550°F., again practically no decarburization from 1550°F. to possibly the A_3 point for pure iron, and above the A_3 point decarburization proceeds with ease. An application of thermodynamics in connection with the iron-carbon equilibrium diagram will reveal that these apparent peculiarities are just what would be expected. The movement of carbon through iron in decarburization being controlled by certain thermodynamic principles leads me to believe that graphitization is a related phenomenon and may be subject to the same principles. Of course, in decarburization the carbon diffuses to the surface and is oxidized to gaseous products, while in graphitization the carbon diffuses to and precipitates on graphite or other particles.

Mr. Timmons has regarded graphitization as a matter of austenite pushing carbon through ferrite to graphite. There are those who would rather start at the other end and think of the precipitation of carbon from ferrite as depleting the concentration in the ferrite, thereby enabling it to pull, so to speak, carbon into itself from the austenite and thus the precipitation of carbon is continued.

In other words, I am thinking of the reaction as leading from the ferrite back to the austenite rather than being pushed from austenite into the ferrite.

CHAIRMAN BOLTON: You were really referring to the rates of reaction in the ultimate end products, the thermodynamics being graphite and ferrite at normal room temperatures. That is what you would have as the background, would you not?

MR. PENNINGTON: That is not only true at room temperature, but at any temperature below the stable eutectoid point. The ultimate products of ferrite and graphite would be the background.

CHAIRMAN BOLTON: But the rate increases very greatly under the range at which autenite is stable. The rate of motion of carbon, migration of carbon, increases rapidly in ferrite as opposed to austenite.

MR. PENNINGTON: Yes, that is true.

The Mechanism of Pin-Hole Formation

By C. E. Sims* and C. A. Zapffe*, Columbus, Ohio

Abstract

The authors have studied the causes of pin-hole formation in steel castings and have reached the conclusion that this phenomenon is caused chiefly by hydrogen. They describe pin-holes and explain the various theories advanced as to their formation, showing why such theories must be discarded. Holding to the theory that porosity is caused by gas precipitated from the steel, the authors explain why the carbon-FeO reaction as the basis for a theory is not tenable and why the hydrogen-FeO reaction affords logical explanation for the formation of pin-holes.

Introduction

1. When steel is cast into sand molds under certain unfavorable conditions, various types of defects or cavities are formed by gas evolution. Such cavities have been designated by various descriptive names such as "blow-holes," "pin-holes," "worm-holes" and by the general term "porosity." Although the various types are produced by different sets of conditions, the nomenclature does not always make as clear a distinction as is desirable in the interests of diagnosis and prevention. The present discussion deals primarily with the type of porosity commonly called "pin-holes," but it is believed that the principle expounded, with some modification, would explain the formation of some other types of porosity.

What Are Pin-holes?

2. Pin-holes are those small, elongated, smooth-walled gas holes which occur immediately under the skin of a casting. The long axis is perpendicular to the surface or parallel to the long axes of the columnar surface crystals. They occur under both vertical and horizontal surfaces and are somewhat more prevalent on concave

* Supervising Metallurgist and Research Metallurgist, respectively, Battelle Memorial Institute.

Note: This paper was presented before a Steel Session at the 45th Annual A.F.A. Convention, New York City, N. Y., May 15, 1941.

surfaces or at re-entrant angles. In the cope face, they may be confused with another type of sub-surface porosity that occurs in connection with inclusions of slag and sand.

3. An outstanding characteristic of pin-holes is that, although they are always very close to the surface, they seldom reach to the surface. There is practically always a layer of sound steel covering them. Sometimes this layer has a substantial thickness, but again it is so thin that it is easily broken or removed by scaling that occurs during heat treating. Many castings, which appear sound when raw, show numerous pin-holes after the heat-treating scale is removed; whereas some require machining to uncover the holes.

4. Other facts well known about pin-holes are: (a) Steels which develop pin-holes when cast into certain green sand molds may be free of pin-holes when cast into well dried molds; (b) A

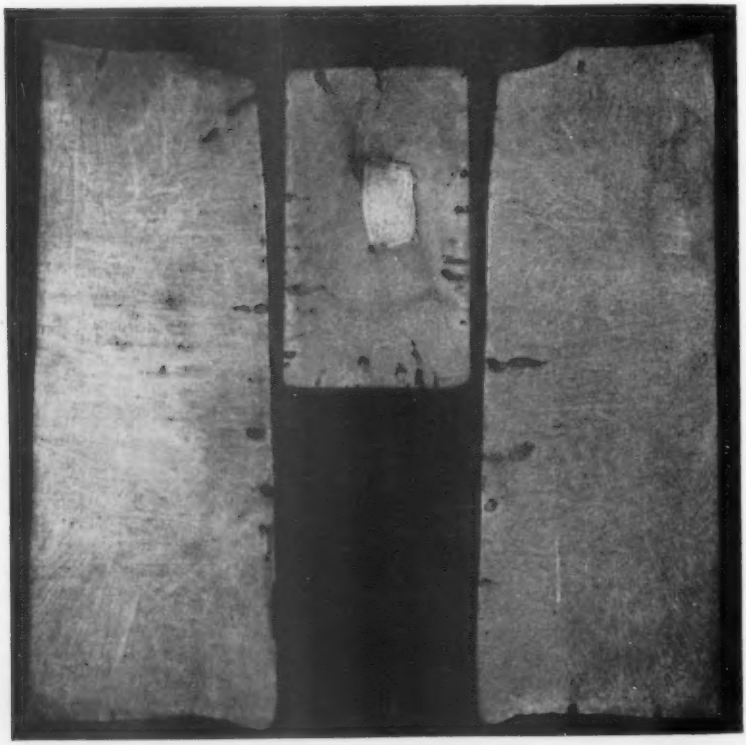


FIG. 1—PIN-HOLE FORMATION IN SEVERAL CASTINGS. THESE CASTINGS HAVE BEEN ANNEALED AND THE PIN-HOLES HAVE BEEN OPENED BY REMOVAL OF SCALE.

green sand mold of high permeability is less likely to cause pin-holes, than a similar mold with low permeability; (c) Steels vary in their propensity for pin-hole formation. A heat of silicon-deoxidized steel, which will have pin-holes when cast in a particular green sand mold, will be quite free from porosity if it is further deoxidized with a strong deoxidizer such as aluminum before casting. Some pin-hole formations are shown in Fig. 1.

Theories on Pin-Hole Formation

5. The majority of expressed opinions inclines to the belief that pin-holes are caused by mold gases forcibly injected into the casting during the very early stages of freezing. Batty[1] attributed them to immigrant mold gases and very ably showed the close relationship of moisture content, permeability, and the resultant pressure developed in the facing sand to the prevalence of pin-holes in the castings. He placed the burden of responsibility on the condition of the mold.

6. Woodward[2] explained the mode of entrance of these gases through the solidified skin of the casting as being afforded by the presence of non-metallic inclusions with low fusion points which, remaining fluid longer than the steel, give way before the pressure of the mold gas and provide gateways through the skin to the molten or plastic interior. He also gives some interesting data to show how the prevalence of pin-holes follows the seasonal variations in humidity. This latter relation indicates that some factor other than the condition of the mold influences the formation of pin-holes.

7. In view of the established relation between gas pressure developed in facing sand and the occurrence of pin-holes, it does not at first seem illogical to attribute these elongated holes to injected mold gas. Furthermore, every foundryman has seen unmistakable blows from moist chills or from improperly dried or poorly vented cores. However, a core blow usually gets under way before the mold is full, and the rising gas bubbles spatter steel out of the gates and risers.

Objections to Theories

8. The first obstacle to the theory of gas injection is the layer of non-porous metal between these holes and the casting surface. Although, in some cases, this wall is so thin that it is removed by

[1] Superior numbers refer to bibliography at end of paper.

scaling, in many cases it has a substantial thickness of obviously sound metal. The slag collar, reported by Woodward[2], is sometimes found, most often in holes under the cope skin; but usually there is no slag in evidence in true pin-holes. Despite some superficial resemblance, it is believed that these holes containing slag and sand which occur so often under the cope skin, sometimes separated from the surface by more than ⅛-in. of sound steel, are different from pin-holes. They vary in size from 1/16-in. to several inches in diameter. One of the present authors[2], in discussing Woodward's paper, advanced an argument to show that included slag particles might be the cause of some types of porosity. It might be observed here that, even after heat treating, those pin-holes not uncovered are bright and unoxidized, which would not be the case if they had even a capillary opening to the air.

9. Even if the skin did have openings through which mold gas, under pressure, could stream, formations like pin-holes would not be produced. As was mentioned earlier, these cavities have rounded cross-sections, smooth walls and the long dimension is perpendicular to the surface, *i.e.* parallel to the columnar surface crystals, whether that surface be vertical or horizontal. The progress of freezing near the surface of a casting is illustrated in Fig. 2. There is a sharp demarcation between the advancing wall of

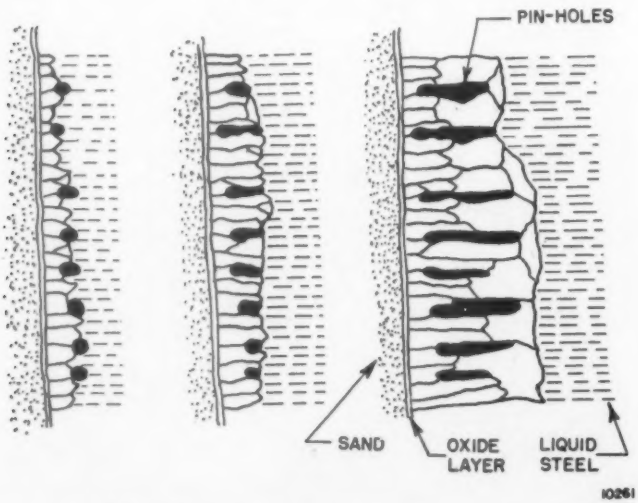


FIG. 2—AUTHORS' CONCEPTION OF THE PROGRESS OF FREEZING AT THE SURFACE OF A CASTING, SHOWING ORIGIN AND GROWTH OF PIN-HOLES.

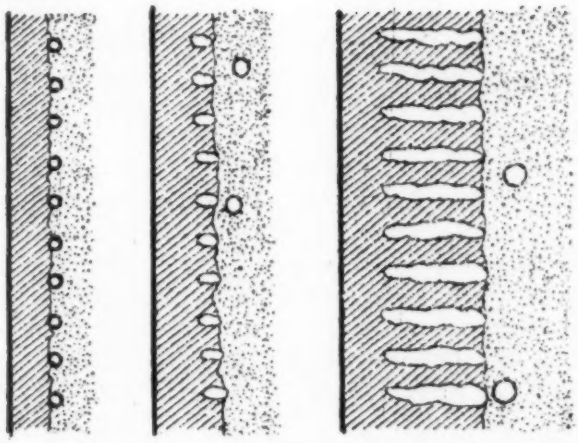


FIG. 3—FORMATION OF PIN-HOLES, ACCORDING TO MÜLLER[3].

solid steel and the liquid, with no intermediate state where crystal dendrites are surrounded by liquid such as occurs nearer the center.

10. The solidified skin is hot, and plastic in the sense of being malleable; but injected gas could not force these crystals apart to form the rounded pin-holes. If the gases enter while the skin is so thin that a hole extends through it, the gases would stream into the liquid and rise as bubbles to the surface. A delicate balance of pressure would be required to force in just enough gas to form a bubble that would cling, especially for forming so many pin-holes as are sometimes found. Also, castings with pin-holes usually do not have an accumulation of gas in the upper parts. The gas injection theory of pin-hole formation appears untenable in view of the characteristics of these cavities. Furthermore, the theory furnishes no mechanism by which aluminum could produce its well-known inhibition of pin-hole formation.

A NEW THEORY

11. A mechanism which appears to fit all the known facts is that the pin-holes are started by bubbles of gas precipitated from the liquid steel adjacent to the frozen skin and that they grow simultaneously with, and in the direction of, the steel crystals in the surface layer of the casting by further precipitation from both the liquid and solid steel. Thus, the bubble constitutes an

inclusion around which the crystals grow. The formation will be discussed in more detail shortly, and evidence will be given that the principal constituent of the gas is hydrogen.

Müller's Theory

12. The idea that pin-holes form in this manner is not new. In 1882 Müller[3] ascribed the formation of pin-holes in steel castings to hydrogen precipitated from the freezing steel and illustrated the mode of formation, as is shown in Fig. 3. He pictures hydrogen dissolved in the molten steel as reaching the saturation point because of concentration on the face of the advancing wall of solid steel by rejection during freezing. It forms first as tiny round bubbles just like the bubbles that form in a glass of water that stands in a warm place. He also argues in refutation of a claim made earlier in the same year by Pourcel[4] that pin-holes are caused by carbon monoxide formed by reaction of FeO and C. Müller's theory was not complete, as will be shown later.

Rim-holes in Rimming Steel

13. The rim-holes of rimming steel ingots have the same shape and similar position in regard to the surface as the pin-holes in castings and unquestionably have the same type of origin and growth. Hayes and Chipman[5] and Hultgren and Phragmén[6], in papers presented contemporaneously, gave lucid and logical descriptions of the manner in which rim-holes or primary blow-holes are formed.

GASES CAUSING PIN-HOLES

14. Since the assertions of Müller and Pourcel, in 1882, up to the present time, arguments have waged continuously as to whether the blow-holes of ingots and castings were formed by CO or H_2. This work has been so well reviewed by Hultgren and Phragmén[6] that no repetition will be made here. In most cases, there seems to have been no clear realization that there might be a difference between the gases produced in a rimming steel and in a killed steel, such as cast steel usually is.

15. Culminating in the work of McCutcheon and Chipman[7] and Hultgren and Phragmén[6], conclusive evidence now shows that the blow-holes of rimming-steel ingots are caused by CO, although there are always small quantities of H_2 and N_2 in the gas evolved,

as would be expected. Swinden and Stevenson[8], on the other hand, gave ample evidence that H_2 can cause porosity in silicon-killed steels and indicated that the oxide/hydrogen reaction plays an important part. It has been shown elsewhere[9] that H_2 bubbled through a silicon-killed steel can cause a very high degree of porosity.

16. Caine[10] concluded that pin-holes in steel castings were caused by evolved gas and that the gas responsible is sometimes H_2 but more often is CO. He maintains that the CO is produced by the reaction $FeO + C \longrightarrow Fe + CO$ during solidification and that this reaction is caused by the ingestion of more FeO than the silicon or other deoxidizers can handle. The reason for the pin-holes being confined to the surface layer is the incorporation of excessive oxides in this layer during pouring.

17. The prohibitive improbability of such a reaction is readily apparent from an examination of the published data on equilibrium constants[11]. Assume, for example, a cast steel containing 0.25 per cent carbon and 0.30 per cent silicon. At a temperature of 2900° F. (1600° C.), both of these elements will be in equilibrium with an FeO content of approximately 0.03 per cent. Just above the freezing point, the FeO would have to be above 0.02 per cent to react with the carbon but only 0.01 per cent FeO can exist in the presence of the 0.3 per cent silicon. The residual silicon acts as a reservoir which will react with any extra FeO that might be introduced, each unit of silicon reacting with 5 times its weight per cent of FeO. To produce a C/FeO reaction, about 0.10 per cent silicon would have to be oxidized from any given portion of the steel, and this would require the introduction of 0.50 per cent FeO to that portion. If it be assumed that such an inordinate quantity of FeO could be absorbed to deplete the silicon in the surface layer, any attempt to explain the action of aluminum is defeated. Residual aluminum would be oxidized in preference to the silicon. As little as 0.01 per cent residual aluminum ordinarily will give protection against pin-holes and it only requires 0.04 per cent FeO to oxidize that much.

18. Nitrogen as a causative agent can readily be eliminated. Chipman and Murphy[12] determined the solubility of N_2 in molten iron to be 0.39 per cent under one atmosphere pressure, and found that the presence of silicon and aluminum did not affect the solu-

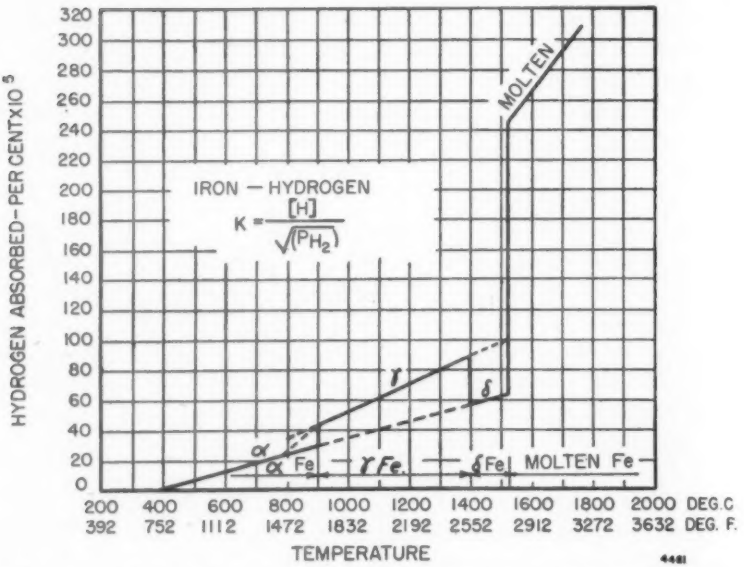


FIG. 4—MODIFIED DIAGRAM OF THE SOLUBILITY OF HYDROGEN UNDER ONE ATMOSPHERE OF H_2, AS DETERMINED BY SIEVERTS.

bility. It also was found that, even when present in the above-mentioned quantity, the N_2 tended to remain in solution rather than to evolve during freezing. The presence of silicon, aluminum, or zirconium under these conditions tended to cause unsoundness rather than to suppress it, probably because oxide inclusions acted as nuclei for precipitation. The quantity of N_2 in an ordinary steel heat never comes within gun-shot of 0.039 per cent; and, in view of these facts, N_2 must be ruled out as an important factor in pin-hole formation. However, it should be realized that, even in the presence of small quantities of N_2, if a bubble of any kind is formed in liquid steel, some N_2 will diffuse into it. The CO given off from a rimming steel always contains small quantities of N_2 and H_2.

HYDROGEN AS A CAUSE OF PIN-HOLES

19. This leaves only a consideration of the factors which indicate hydrogen to be the cause of pin-holes and the mechanism by which pin-holes are produced. As is well known, hydrogen dissolves atomically in both solid and liquid iron. The solubility under one atmosphere of H_2, as determined by Sieverts, is shown in a modified diagram in Fig. 4. Sieverts also determined that the sol-

ubility varies as the square root of the pressure of H_2. The relation between pressure and solubility is shown graphically in Fig. 5.

20. It will be observed from Fig. 4, that there is a rapid decrease in solubility with temperature above the freezing point, an abrupt drop at the freezing point, and a continued decrease below it. The various methods by which hydrogen may be introduced into or eliminated from steel have been discussed at length in the literature and will not be reviewed at this time.

21. To avoid a common error in interpreting the implications of Sieverts' curve for hydrogen solubility, it should be borne in mind that it represents conditions only for one atmosphere pressure of H_2. With any other pressure of H_2, the curve would shift either up or down. Hydrogen does not have a definite solubility at a given temperature, such as carbon has, for example. With a zero partial pressure of H_2 in the atmosphere in contact with a piece of steel, that steel is always saturated with hydrogen regardless of its content.

22. Because only atomic hydrogen dissolves in steel, the dissociated portion of the hydrogen atmosphere controls the solubility. The dissociation increases markedly with temperature, and this

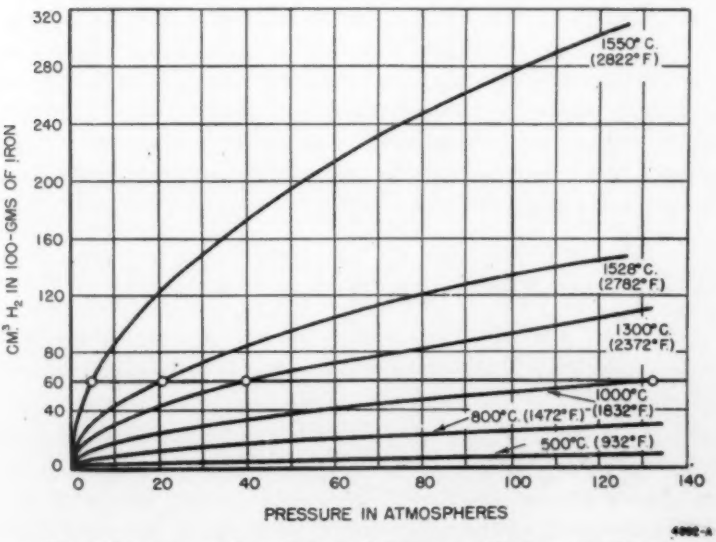


FIG. 5—RELATION OF SOLUBILITY TO PRESSURE FOR HYDROGEN IN STEEL (BARDENHEUER AND KELLER).

explains the relation between solubility and temperature. This also means that, if one atmosphere of H_2 at a given temperature provides a concentration of 10 atoms per unit area on steel, and water vapor by the reaction

$$Fe + H_2O \longrightarrow FeO + 2H$$

provides a concentration of 50 atoms for the same area, the hydrogen dissolved because of the water vapor will be equivalent to that from *25 atmospheres* of H_2 (square root relation between pressure of H_2 and solubility of H).

Solubility of Hydrogen and Nitrogen

23. One belief, that has been popular and often expressed, is that deoxidizers apparently increase the solubility of nitrogen and hydrogen in steel and thus prevent porosity. Chipman and Murphy[12] showed that this is not true for nitrogen and there is no apparent evidence to indicate that it is true for hydrogen. Such a theory, furthermore, is entirely unnecessary to explain the phenomenon of porosity.

24. Both of these elements do have a strong tendency to form solid solutions rather than to precipitate as bubbles during freezing of the metal in which they are dissolved. This was shown for N_2 by Chipman and Murphy[12] and "it has been demonstrated that liquid copper, containing comparatively large quantities of hyrogen, can be cast into ingots without the formation of blow-holes, provided that the necessary precautions are taken to prevent the metal from becoming contaminated with oxygen, either before or during casting. If, however, the same metal is allowed to absorb oxygen during casting, a very large number of blow-holes are formed in the ingot."[13]

EXPERIMENTS TO TEST THEORY

25. Steel behaves in a similar manner. In some experimental work at Battelle Memorial Institute small heats of steel were melted, cast into ingots, and allowed to cool, all under one atmosphere of pure hydrogen. High-frequency induction-heating was used for melting. Ample opportunity was allowed for the molten steel to become saturated with hydrogen; but, if the ingot mold was scrupulously cleaned of scale or other oxide, the ingots came out silvery white, deeply piped, and entirely free of blow-holes.

If, on the other hand, an ingot mold with some adhering scale or other oxide was used, the ingots literally boiled out of the mold and were much too porous to use.

26. It must be concluded, therefore, that hydrogen dissolved in steel, even to saturation under one atmosphere of H_2, is unable to initiate bubble formation. If a bubble of any other material is introduced under such conditions, hydrogen will diffuse into it and cause it to grow with great rapidity. The most likely source of such a bubble is water vapor formed by the reaction

$$2H + FeO \longrightarrow Fe + H_2O.$$

27. The reason for this assumption becomes apparent with a study of the pressure-solubility relation shown in Fig. 5. It has been shown that hydrogen has no definite solubility in steel; the solubility is entirely relative, dependent on the external pressure of hydrogen. The condition of saturation, therefore, can exist only at a surface where the dissolved hydrogen can be in equilibrium with the hydrogen in the atmosphere. Saturation or supersaturation cannot be attained within the body of liquid steel and thus bubbles cannot start as H_2.

28. The concentration of dissolved hydrogen, however, has a direct equilibrium relation with FeO; and, whenever the product of $[H]^2 \times [FeO]$* exceeds the equilibrium constant K, the reaction to form water vapor will occur. This water vapor will form a bubble which will constitute an interior surface and hydrogen will precipitate into it to establish the pressure-solubility relations of Fig. 5. As long as the bubble is in contact with liquid steel, it can grow; and the total pressure, which is limited by the ferrostatic head, cannot get very much above one atmosphere.

Hydrogen in "Flat" Melted Steel

29. The constant

$$K = [H]^2 \times [FeO]$$

for the reaction is unknown and it is not yet possible to calculate the relative concentrations of H and FeO that will react. However, some relations are known qualitatively. Steel melted flat, *i.e.* without any boil, often will contain so much hydrogen that 0.30 to 0.40 per cent silicon will not lower the FeO sufficiently to prevent the reaction to form H_2O and the steel will freeze porous. In this

* Square brackets denote solution in iron.

case, the porosity will be general throughout the cross section, as in a steel insufficiently deoxidized to prevent the C/FeO reaction. Aluminum added in sufficient quantity to such a steel, will prevent porosity by lowering the FeO below the point of equilibrium with the hydrogen present.

Hydrogen in "Boiled" Steel

30. On the other hand, steel properly boiled to lower the hydrogen content, and subsequently protected from undue absorption, will freeze quite free of porosity if deoxidized with 0.3 to 0.4 per cent silicon. In this case, the hydrogen is too low to react with the FeO in equilibrium with the silicon. Such a steel, cast into a well-dried sand mold, will have no pin-holes.

Action in Green Sand Mold

31. Should this steel be poured into a green sand mold, the following sequence of events would occur: Almost immediately when the molten steel contacts the cold, moist sand, a thin skin of solid steel forms. At the same time, the water in the adjacent sand is changed to steam with an increase of some 5000 volumes and creates an atmosphere highly oxidizing to steel. A portion of this water vapor escapes through the porous sand, but some of it reacts with the hot steel thus:

$$H_2O + Fe \longrightarrow 2H + FeO$$

32. FeO is formed as a layer of scale on the surface, where it stays because of its very slow rate of diffusion. It may be further oxidized to Fe_3O_4, but for all practical purposes it does not affect the freezing steel.

BUBBLE GROWTH

33. The atomic hydrogen, released in contact with the steel, readily dissolves; and the rate of diffusion is so great at temperatures just under the freezing point that it flows through the thin layer of solid steel about as readily as the water vapor penetrates the facing sand. In this manner, a high concentration of hydrogen is built up in the liquid steel in contact with the solidified skin.

34. When the concentration becomes high enough, hydrogen will react with dissolved FeO to form a small bubble of water

vapor on the frozen wall. Immediately hydrogen from the adjacent liquid steel, as also that still diffusing through the solid layer, will begin diffusing into this bubble to form gaseous molecular hydrogen which will not dissociate sufficiently to redissolve until equilibrium with remaining dissolved gas is established. Thus the bubble will grow.

35. Hydrogen segregates during freezing, the same as any other substance dissolved in steel. This was demonstrated experimentally. Two low-alloy steel melts were made, identical in composition except that one was made in vacuo and the other in hydro-

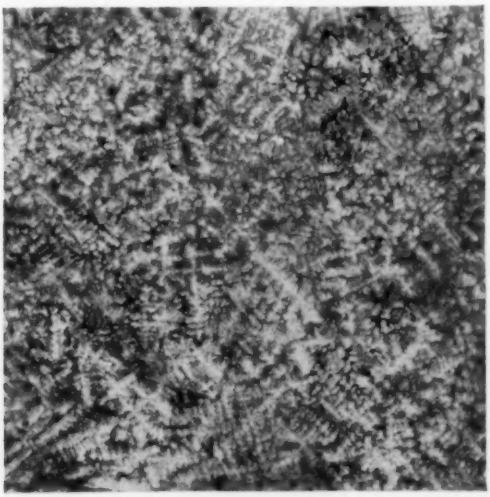


FIG. 6—INTERDENDRITIC VOIDS CAUSED BY DENDRITIC SEGREGATION OF HYDROGEN ABSORBED DURING MELTING, 5X.

gen under one atmosphere pressure. The hydrogen-containing ingot had no blow-holes; but, as shown in Fig. 6, it had numerous voids throughout the interdentritic interstices caused by the hydrogen which had precipitated there. The hydrogen-free ingot showed no such interdendritic voids even after deep etching. There probably always will be some concentration of hydrogen, therefore, adjacent to the advancing wall of freezing steel; and this will aid bubble growth, once a bubble has started.

36. While the bubble is growing, freezing is also going on; and the columnar surface crystals are growing rapidly. The crystals grow around the bubble, leaving it in a recess, but do not readily

close over it because the bubble acts as an insulator through which the latent heat of crystallization cannot be dissipated. The bubble and the crystals thus grow simultaneously; if the bubble grows too fast, portions will become detached and float off, so it is constrained to grow as a narrow channel, the diameter of which cannot exceed that of the largest attached bubble. When the growth of the advancing bubble slows down sufficiently because of depletion of hydrogen in the surrounding metal, the crystals seal off the passageway and a pin-hole remains.

37. Numerous bubbles of water vapor are formed simultaneously to start as many pin-holes; but from that time on, the decrease in concentration of the hydrogen, by collapse into the bubbles, effectively prevents any new bubbles of water vapor from forming. For this reason, there is only one row of pin-holes and the steel from there on is sound. Some nitrogen will always be found in the gas of pin-holes.

EFFECT OF ALUMINUM ADDITIONS

38. When aluminum, in sufficient quantity to leave a small residual content of the free metal, is added to a silicon-killed steel, the FeO content is lowered to such an extent that even the influx of hydrogen from moist sand cannot raise the concentration high enough to cause a reaction to produce water vapor. Under this condition, no pin-holes are formed, although the hydrogen content actually may be higher. It was shown 30 years ago[14] that sound steel often contains more hydrogen than unsound steel.

EFFECT OF MOLD ON POROSITY

39. Other things being equal, the initial hydrogen content of a steel will influence its tendency to form pin-holes. With a given steel, such factors as moisture content and permeability of the facing sand and contour of the casting will determine the prevalence of pin-holes. That is to say, any condition which will build up a high pressure of water vapor on the face of the casting will set the stage for a large flow of hydrogen into the casting.

DIFFERENCES IN REACTIONS

40. There is a fundamental difference in the consideration of the C/FeO reaction or the H/FeO reaction as the origin of pin-

holes. With carbon invariant and with a normal silicon content of, say 0.35 per cent, the former reaction could take place only through a large influx of FeO, enough to deplete the back-log of residual silicon. In this case, aluminum, because of the usual small reserve, would be powerless to influence a condition not controlled readily by the silicon.

41. With the second reaction, the FeO remains constant as determined by the silicon or aluminum (there must be at least a small residual aluminum to insure efficient deoxidation), and the hydrogen varies by the means described. The steel has no protection against increase in hydrogen concentration as it has against FeO increments, and if the hydrogen gets above its equilibrium value with FeO, there will be a reaction to form H_2O. Aluminum, by decreasing the FeO almost to the vanishing point, greatly increases the hydrogen concentration necessary to start a reaction. It is believed that a similar mechanism works in other types of porosity which aluminum is effective in preventing.

SUMMARY

42. As a result of their investigations, the authors have reached the following conclusions:

1. Pin-holes are not caused by forcible injection of mold gases through the skin of a freezing casting.

2. They are caused by precipitation of gas from solution in the steel.

3. The H/FeO reaction to form H_2O is the origin of pin-holes rather than CO from the C/FeO reaction.

4. After their inception, pin-holes grow by H_2 precipitating from solution.

5. Mold gases, principally water vapor, furnish the excess of H_2 that starts pin-hole formation.

Bibliography

1. Batty, G., *"The Most Potent Variable,"* TRANSACTIONS, AMERICAN FOUNDRYMEN'S ASSOCIATION, vol. 38, pp. 309-331 (1930).

2. Woodward, R. C., *"The Mechanics of Porosity in Steel Castings,"* IBID., vol. 42, pp. 364-374 (1934).

3. Müller, F. C. G., *"Ueber die Gasausscheidungen in Stahlgüssen,"* STAHL UND EISEN, vol. 2, pp. 531-542 (1882).

4. Pourcel, A., *"Notes on the Manufacture of Solid Steel Castings,"* JOURNAL IRON AND STEEL INSTITUTE, vol. 22, pp. 509-533 (1882).

5. Hayes, A., and Chipman, J., *"Mechanism of Solidification and Segregation in a Low-Carbon Rimming Steel Ingot,"* TRANSACTIONS, AMERICAN INSTITUTE OF MINING AND METALLURGICAL ENGINEERS, vol. 135, pp. 85-132 (1939).

6. Hultgren, A., and Phragmén, G., *"Solidification of Rimming-Steel Ingots,* IBID., vol. 135, pp. 133-244 (1939).

7. McCutcheon, K. C., and Chipman, J., *"Evolution of Gases from Rimming-Steel Ingots,"* IBID., vol. 131, pp. 206-230 (1939).

8. Swinden, T., and Stevenson, W. W., *"Some Experiments on Gases in Iron and Steel and Their Effect on the Solidification of Ingots,"* Iron & Steel Institute, *"Sixth Report on Heterogeneity of Steel Ingots,"* pp. 137-150 (1935).

9. Sims, C. E., *"Preparation of Steel to Avoid Porosity in Castings,"* TRANSACTIONS, AMERICAN FOUNDRYMEN'S ASSOCIATION, vol. 42, pp. 323-338 (1934).

10. Caine, J. B., *"A Method of Control of Oxides and Gases in Steel for Castings,"* Unpublished paper presented at Round Table discussion, American Foundrymen's Association Convention, May, 1940.

11. Chipman, J., *"Application of Thermodynamics to the Deoxidation of Liquid Steel,"* TRANSACTIONS, AMERICAN SOCIETY FOR METALS, vol. 22, pp. 385-446 (1934).

12. Chipman, J., and Murphy, D. W., *"Solubility of Nitrogen in Liquid Iron,"* METALS TECHNOLOGY, vol. 2, January, Tech. Publ. 591, 12 pages (1935).

13. *"Principles Involved in the Making of Rimming Steel,"* Iron & Steel Inst., *"Fourth Report on the Heterogeneity of Steel Ingots,"* pp. 68-99 (See page 88) (1932).

14. Baker, T., *"Gases Occluded in Steel,"* Iron and Steel Institute, CARNEGIE SCHOLARSHIP MEMOIRS, vol. 3, pp. 249-259 (1911).

DISCUSSION

Presiding: JOHN HOWE HALL, Philadelphia, Pa.

J. B. CAINE[1] *(Written Discussion):* The authors' objection, as stated in paragraph 17, to the possibility of pin-hole formation by CO gas in silicon killed steels, based on the equilibria in molten steel just above the freezing point, is perfectly valid. It is true that at this point the Si controls the FeO content and that any reaction occurring will form silicate inclusions and not CO gas. The conditions of pin-hole formation point unmistakably to the fact that the rejection of gas occurs during solidification of the steel. The authors concur with this statement in their conception of the origin of pin-holes in Fig. 2. During solidification, however, we have an entirely different set of equilibria than we had just above the solidification temperature, because of segregation occurring during solidification. The amount of segregation of

[1] Metallurgist, Sawbrook Steel Castings Co., Lockland, O.

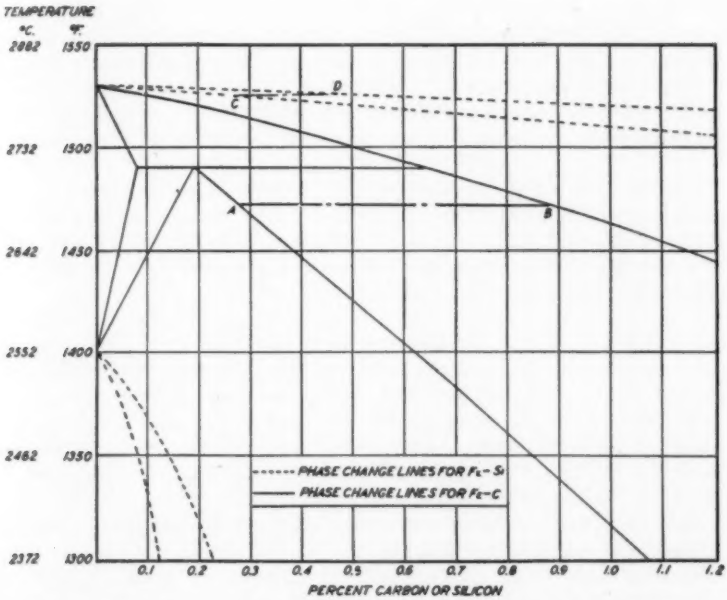


FIG. 7—CHART ILLUSTRATING THE SEGREGATION OF CARBON AND SILICON DURING SOLIDIFICATION.

C and Si during solidification is shown in Fig. 7. of this discussion* †, showing the portion of the phase diagrams in which we are interested. The two diagrams have been plotted together to show the differences, and should not be taken as one phase diagram.

By applying the lever law, it is seen that when the metal is partly solid and approaching the solidus line, the C in the remaining liquid has increased from 0.30 per cent to about 0.90 per cent (point B Fig. 7). The Si at the same point, because of the much smaller gap between the solidus and liquidus lines, has only increased from 0.30 per cent to about 0.45 per cent, (point D). Therefore, because of the much greater segregation of C during freezing, we have just the opposite condition we had just above the solidification temperature. Instead of 0.02 per cent FeO in equilibrium with 0.30 per cent C, 0.009 per cent FeO in equilibrium with 0.30 per cent Si in the liquid metal, and the formation of silicate inclusions, we have 0.006 per cent FeO in equilibrium with 0.90 per cent C and a higher FeO content, 0.0075 per cent in equilibrium with 0.45 per cent Si during the latter part of solidification. The formation of CO gas without the oxidation of a large amount of Si is now possible. As the equilibrium contents of C and Si, with regard to FeO, are so close together during solidification, just which reaction product will be formed will depend on the exact temperature and concentration gradient

* Epstein, Alloys of Iron and Carbon, vol. 1, p. 96.
† Greiner, Marsh and Stoughton, Alloys of Iron and Silicon, p. 51.

at each liquid solid interface. At temperatures just below the liquidus line, silicate inclusions will form, but as the temperature approaches the solidus line and the C content of the remaining liquid has increased sufficiently, CO gas is formed. So what we actually find is a simultaneous formation of silicate inclusions and CO gas in the mass. This point has been checked repeatedly by a study of the inclusions.

In the case of H porosity, the FeO content can be lowered to extremely low amounts without eliminating the pin-holes. If a heat is "melted flat" without a boil, neither Al, Mg, Ca or Zr will eliminate the rejection of H, either by the FeO-H reaction or straight H solubility and subsequent rejection during solidification.

In paragraph 32, the authors state that the FeO formed by the reaction between the steel and the water in the sand does not diffuse into the metal. How else can they explain the formation of innumerable oxide inclusions in the steel next to the sand metal interface when it is poured in sand containing water, other than the absorption of oxygen, above and during solidification. In extreme cases, such as illustrated in Fig. 5b of my report[10] before the A. F. A. last year, there has been enough oxygen absorbed to not only oxidize all the Al and Si present at the surface of the steel, but also the Mn, as the sulphide inclusions in these areas are predominately FeS. This surface oxidation and decarburization and its dependence on the water content of the sand has been checked in hundreds of specimens from quite a number of steel foundries in the middlewest.

The FeO-H reaction advanced by the authors is valuable in not only explaining some types of porosity, but should prove extremely important in explaining flake formation in the solid metal. It is my opinion, at least, that it should be advanced as one of a number of causes of porosity in steel castings, not the only one.

MESSRS. ZAPFFE AND SIMS *(Reply to Written Discussion):* Caine's statement that the FeO, formed by the water in the sand, penetrates the steel to oxidize certain elements in the steel is entirely correct and is well known. The action of mold moisture on hot steel is simultaneously oxidizing and hydrogenizing.

That oxygen injection cannot be an important factor in causing pin-holes, however, follows from simple reasoning. First, water decomposes to liberate twice as much hydrogen as oxygen. Secondly, whereas the oxygen, except for a minute portion that remains atomically dissolved, immediately combines both with the elements in iron and with the iron itself to form stable, non-diffusing, solid compounds, hydrogen behaves simply as a gas in solid solution, with the exception of its partial re-combination within the metal to form the H_2O that causes pin-holes. Thirdly, the diffusion rate of hydrogen in the hot steel is many hundreds of times greater than that of atomically dissolved oxygen. In

[10] See reference number in Sims and Zapffe's bibliography.

the split-second time factor of pin-hole formation, only hydrogen could diffuse fast enough to play a part. True, oxide can be found at and near the surface of the casting, for the water vapor from the mold would not decompose unless the metal reacted with it, and the oxide is the reaction product that attests that decomposition. The absorption of oxygen is too slow and shallow, however, to affect pin-hole formation.

There is little doubt that on rapidly cooling some steels a C-O reaction might proceed in order to regain the equilibrium upset by the rapid cooling. Caine's argument, however, is defective in several important respects. First, the system Fe-Si-C is ternary and cannot be expressed by two binary diagrams. Secondly, Caine has neglected to consider that the control of FeO by C is a linear relationship, whereas the *square root* of Si controls FeO, in accordance with the common steel-making reactions:

$$FeO + C = Fe + CO$$
$$2\ FeO + Si = 2\ Fe + SiO_2$$

To use Caine's own figures, the concentration of carbon increasing from 0.30 to 0.90 per cent, which is a threefold increase, should be accompanied by a $\sqrt{3}$ increase in Si. Again using Caine's figures, one can see that Si should increase from 0.30 per cent to (1.7 x 0.30 per cent) = 0.51 per cent Si, which is as close an approximation to the actual value as can possibly be gained from the slowly sloping line on the equilibrium diagram. The enrichment of the liquid shown on constitution diagrams is in itself an equilibrium reaction and cannot *per se* be used to explain an upset in equilibrium. The enrichment occurs simply because the so-called vapor pressure of the solute is less in the liquid than in the solid. An analogy can be drawn with alcohol and benzine in an atmosphere of water vapor. Under the same pressure of H_2O, much greater enrichment will occur in the alcohol with respect to the absorption of H_2O in the benzine.

The chemical potentials of the various elements, which are proportional to these vapor pressures, are the only thermodynamic entities that determine whether a reaction will proceed. They may very often have but little resemblance to actual concentration and follow regular fundamental relationships that can be expressed by the van't Hoff Isochore:

$$\frac{d \ln K}{dT} = \frac{H}{RT^2}$$

Those relationships are fairly well established for steel-making reactions and show that the respective equilibria of Si-FeO and C-FeO have temperature coefficients such that with decreasing temperature the Si-FeO reaction is favored. Si will reduce CO upon cooling just as C will reduce SiO_2 upon heating. In the rapid cooling necessary for retaining the constitutional enrichment as a non-equilibrium segregation, the enrichment never reaches the values shown in Fig. 7. Again, by the time the enrichment has progressed very far, only minute portions of liquid remain.

Caine infers that a steel containing 0.90 per cent C and deoxidized with 0.50 per cent Si cannot be solidified without the evolution of gas from the C-O reaction. Experience teaches us that the contrary is true.

The observation that some steels which are melted "flat" and consequently loaded with hydrogen, will be gassy even though deoxidized with aluminum or some other strong deoxidizer, is undoubtedly true. It is possible that if the hydrogen content is high enough, even aluminum may not carry the FeO low enough to prevent reaction. We do not have personal knowledge of such an instance. It should be added, however, that our personal experience does include some instances where it was concluded that Al would not stop pin-hole porosity, and investigation showed that the quantity of Al added was insufficient for complete deoxidation. Unless there is some residual deoxidizer after deoxidation, there is no assurance of efficient deoxidation.

We cannot agree with the statement that the FeO-H reaction may cause flaking. Flakes that are formed when a steel is fractured are caused by hydrogen concentrations alone, and shatter cracks are caused by a combination of hydrogen concentration and stress.

It probably should be reiterated here that the present paper attempts only to explain the mechanism of pin-hole formation and not porosity in general.

H. F. TAYLOR[2]: It occurs to me that we may just possibly be looking for too hard an explanation for the initial formation of pin-holes; but regarding how to prevent them, I think that Messrs. Sims and Zapffe have given the solution. I think they have provided the industry with a mighty fine tool. Mr. Sims mentioned a similarity between pin-holes and the holes that would occur on the cope surface of a casting and, in one instance, he mentioned a particle of slag which possibly extended to the surface. He also mentioned that the cavity formed around this particle of slag was several inches in diameter. Is that right, Mr. Sims?

MR. SIMS: In some cases holes are that big, but in those the slag does not extend to the surface.

MR. TAYLOR: It does not seem possible for such a small amount of hydrogen, as we have been shown is evolved, to form a hole several inches in diameter. I wonder if the explanation does not lie somewhere else? We know that the ferrostatic pressure will probably be a minimum on the cope surface of the casting. We also have had brought to our attention the possibility of atmospheric pressure coming into play in the feeding of castings. We know that in internal angles, where there are hot spots, we find a large number of these cavities called pin-holes. Is that right, Mr. Sims? Do you find a preponderance of pin-holes in internal angles or hot spots?

MR. SIMS: Yes.

---

[2] Naval Research Lab., Anacostia Sta., Washington, D. C.

MR. TAYLOR: Then may I venture the suggestion that atmospheric pressure might cause the hydrogen to penetrate into the thin skin of the casting. This would be in excess of such pressure as might be generated by the evolution of hydrogen through the reaction of the water vapor and the iron and the subsequent reaction of iron oxide.

Mr. Sims showed us columnar crystals forming around the little black cavities that proved so troublesome. It seems to me that rather than solution into and subsequent rejection of the hydrogen from the metal, perhaps the extra pressure has caused the hydrogen to force its way in between the crystals or at least to put the hydrogen in there far enough to cause the reaction. By having the aluminum present and lowering the iron oxide to perhaps one-tenth of what it would be with silicon, Messrs. Sims and Zapffe have made it impossible for the hydrogen to do any damage. Thus I wonder if the formation of pin-holes is not tied up with pressure which may be equal to or in excess of ordinary atmospheric pressure?

I want to clarify my stand regarding the slag particle. The other day we had occasion to make a casting, and this was a very special casting. We have made lots of them satisfactorily but this particular time, men from the Carnegie Institute of Technology were witnessing the test and, like all things when you are trying to show off, it went sour. Just inside the test casting we noticed a large cavity. These cavities are not unknown to us. It was not the first time we had run into it, but it was an exception. We sawed the casting apart and just under the cope surface was a little round hole. This hole barely reached the outside. To me it means that the atmospheric pressure forced itself into the casting, which was incompletely fed, and this was made possible by the slag creating a hot spot in the cope surface of the casting.

MR. SIMS: Was that last hole exposed to the surface?

MR. TAYLOR: Yes, it was.

MR. SIMS: It did not have the skin over it?

MR. TAYLOR: No, it was a little hole about as large as a small needle.

MR. SIMS: It was my purpose to differentiate between pin-holes and the holes under the cope surface and point out the fact that I was not including the latter, but was speaking specifically of pin-holes. These large holes, containing sand or slag, have a different genesis, although I think there are some points of similarity.

The pressure of water vapor on the surface is unquestionably a factor of importance. The greater prevalence of pin-holes in re-entrant angles is undoubtedly caused by the lowered opportunity for the escape of water vapor out through the sand and the consequent higher pressure that is built up at such places. The higher pressure, of course, provides greater concentration of atomic hydrogen at the surface, which results

in greater and more rapid diffusion of the gas into the metal. But pressures sufficient to force gas in between the crystals would likewise blow the sand away from the casting in my opinion.

I doubt that enough pressure could be developed on the surface of a casting to actually force gas through a solidified face. Furthermore, such a mechanical action is unnecessary for explaining pin-holes when the simpler picture of diffusion can be used. Again, if the gases, which are originally water vapor, were blown in mechanically, the surface of the pin-holes would be oxidized. This is not true.

MR. TAYLOR: I did not clarify one other point. The reason why a preponderance of pin-holes occurs at the internal angle is that we have a hot spot which creates a weak point in the casting skin. I am sure we have all seen the little draw or possibly the hole that reaches to the surface, on the narrow section with the square corner, for example, or in the legs of a "Y" section, where we have a sharp wedge of sand. Everywhere else atmospheric pressure is acting fine but it is not able to exceed the ability of the steel to push back against it. But at this hot spot, it is perfectly possible, I am sure since I have seen it in so many instances, for a hole to blow through, maybe a hole so small you can hardly see it. I had to have a microscope one day to find a hole I knew was there.

Now it seems to be possible, that in line with the fact, that there are holes so near the surface, and in a good many cases actually do come to the surface, that atmospheric pressure does come into play and that infinitely small hot spots may be initiated by small inclusions or in other ways, so that the intercrystalling bond is weak at elevated temperatures.

I just wanted to point out that the hot spot does not necessarily increase the pressure. It is true, however, that if it happens to be a green sand mold or one containing gas forming binders, it will give off excessive mold gas. If a pronounced hot spot exists, atmospheric pressure alone is ample to do the trick.

H. D. PHILLIPS[3]: Mr. Taylor brings up a very good point, although I believe it is slightly different from the type of porosity which Mr. Sims has endeavored to explain to us in his paper. There have been many theories advanced on the mechanism of the formation of this pin-hole type of porosity in steel castings. Some have a reasonable foundation in fact, but they have never satisfactorily explained this trouble to most of us.

Sims and Zapffe have given us the results of some real thought and have advanced in this paper, what appears to me to be, a logical explanation of the formation of this pin-hole type porosity. The industry as a whole, and particularly the acid electric steel foundry people, are indebted to the authors in this case, and to Mr. Sims and his associates

[3] Plant Mgr., Lebanon Steel Foundry, Lebanon, Pa.

in other cases. First, in 1932, I believe it was, before the A. I. M. E., he gave us a real, logical, clear explanation of why aluminum causes poor ductility in steel castings. We used aluminum to prevent porosity, but we had the attendant problem of poor ductility. Then he came along and told us how we could use aluminum, which would eliminate our porosity and at the same time, retain our good ductility to a large extent.

In this case he is giving us a logical theory which we can use, and it seems consistent with all the facts and observations I have made. He has given us a logical theory which we can use in attacking this problem and solving it. If we have a logical theory which is consistent with all observed facts, we have gone a long way towards finding the cure.

What is significant to me in this paper is that he divides the responsibility between the melter and molder, or between the mold and the metal. He does not blame either, it is a case of mutual responsibility in the formation of this type of defect.

One thing that appears to be explained by his theory is the fact that porosity of this type is more prevalent in light steel castings; that is, steel castings of light metal section. I would say, assuming two castings where every dimension is the same excepting thickness, the casting that would have ¼-in. metal section with re-entrant angles, that porosity would be more evident in that casting than in the other casting with a metal section of 1-in. While such things as temperature and other items enter into the picture, I think it is a case of hydrogen solubility and mass.

Why do we find, in our experience, that porosity of this type is less evident in the austenitic type steel than in the ordinary carbon steel? We find that it is not necessary to use aluminum in castings of austenitic steel, whereas, the same type casting, in carbon steel, will require aluminum. Fig. 5 gives an explanation of this.

JOHN JUPPENLATZ[4]: Mr. Sims has brought out some valuable information for all foundries, especially those producing green sand castings. While he has stated in his paper that hydrogen is a primary cause of pin-hole formation, I do not know as I can fully subscribe to the theory that when water dissociates two volumes of hydrogen and one volume of oxygen forms. Molten steel has an affinity for both of these products. He has mentioned in his paper where the FeO and CO reactions take place. I think sometimes there is a combination of these ills in connection with hydrogen, as he has shown. In actual practice, in melting acid steels in the electric furnace, melting them very dead, with no appreciable boil and possibly a high recovery of residual silicon and manganese, I have found it very difficult to deoxidize or so fix the heat of steel so that pin-hole formation could be eliminated.

[4] Metallurgist, Treadwell Engineering Co., Easton, Pa.

Mr. Sims has brought this point out and I think it is not overemphasized when I state that pin-hole formation due to hydrogen content of the metal is a very important factor. Efforts during melting should be made to drag the hydrogen content of the steel to the lowest possible limits. Aluminum, in high hydrogen content of molten steel, does not seem to completely prevent the pin-holes, either in green or dry sand. When I speak of that, I could say that the aluminum content probably added in this connection in the stream might have gone to 5 lb. per ton.

Sometime ago, we had small pin-hole formations in some rather light castings. The pin-holes occurred in the green sand molds, not in the dry sand molds, and only on the drag side of the casting. I do not think we ever had a good explanation of why this condition occurred. This condition of small pin-holes on the drag side went away as quickly as it came and I believe if we had investigated further, we would probably have found the condition in the sand, possibly iron oxide or dust of that nature in the sand, which was sufficient to react with the moisture on the drag side of the mold. Possibly the heat of the molten metal would drive a thin surface of the moisture off the cope side of the casting, thereby leaving the practically moisture free green cope. I would like to hear if Mr. Sims has any reactions as to this peculiar condition?

MR. SIMS: Mr. Phillips' discussion brings up a subject that I did not mention. There are quite a few aspects of the hydrogen problem that were not brought out, either because they were not particularly pertinent to the present paper or because there was a lack of time. One is the difference in the porosity of a ferritic steel; that is, a steel that freezes as ferrite such as a high chromium steel. The ferritic steels have a tendency to be porous and in ingot practice they produce bleeding ingots. This bleeding is apparently caused by the difference between the solubility of hydrogen in gamma and in alpha iron.

As was mentioned in reply to Caine, Mr. Juppenlatz is probably correct in his statement that if enough hydrogen is dissolved in steel, such as by melting flat, even as strong a deoxidizer as aluminum may not prevent evolution of gas during freezing. The removal of some hydrogen by boiling is of great importance. Interpretations of data from Al added to the mold are open to question. There is a considerable burden of responsibility on the operator to get uniform distribution and if he is a little late in getting started with his addition, the metal of the drag surface may be deficient. There also is a time element in getting distribution of Al in steel and distribution is sometimes incomplete when added to a hand shank, and more so when added to a stream entering a mold. Individual cases should be judged only on a basis of all the facts, but it seems probable that the case Mr. Juppenlatz mentioned was due to a moisture-permeability combination that was wrong for the green sand.

F. A. MELMOTH[5]: The hypothetical reasonings of papers of this type either stand or fall to the degree in which they fit, or fail to fit, the facts, as shown in practice.

I am interested, however, as are most foundrymen, in just how much can we get out of these discussions to overcome practical trouble, and that is where I think we owe a debt of gratitude to Mr. Sims, because, although he is capable of giving the highly theoretical hypotheses, he always gives a guide to where they can be applied fairly easily. What I would like to do is to take the main conditions as we know them that appear to produce this type of pin-hole porosity and just parallel his hypotheses with those conditions and see how they fit together.

First, there is the metal. I think everybody has agreed at various meetings in discussing metal that we have arrived at the point where the boil is admitted to be of prime importance. Mr. Sims, together with other contributors, have proved to us that this condition had a sound basis. We are lowering gas concentration and the boil is apparently the only way where certain gases are concerned. Therefore, his hypothesis of gas concentration and hydrogen probably can be said to fit to the production of the type of porosity occurring in the absence of a satisfactory boil. There is no doubt whatever in my mind, and has not been for many years, regardless of any suitable scientific explanation, that if we did not boil our steel, it was liable to pin-hole porosity.

Now the next practical point would be imperfectly dried ladles. We have tracked down cases where ladles have been used imperfectly dried and this type of porosity has resulted. Again, we subjected our steel to the possibility of increased gas absorption. The hypothesis fits.

Then take the question of the molding sand permeability. I am referring now to facing. If the tendency to produce impermeable mold faces is greater, the possibility of building up more pressure is greater during pouring, and again the steel is subjected to increased gaseous pressure at a time when it can ill afford to stand it. The hypothesis again fits.

Then there is the question of the backing sand. I am convinced a lot of porosity has been caused by incorrect balance between the permeability of the backing sand and the facing sand, when the mechanical action probably was gas pressure set up behind the facing which reversed itself and turned towards the mold face. In that case one is once again building up mold pressure and subjecting the steel to increased possibilities of gas absorption. There again the hypothesis fits. The only place I have noticed where the hypothesis might not fit is, in the case describable as pin-hole porosity, occurring at radii. I do not know whether the cases are truly parallel.

We can imagine a film forming on the face of a mold during pouring and on either side by any angle the sand has a certain capacity for

[5] Vice President, Detroit Steel Casting Co., Detroit, Mich.

removing heat from the steel and producing a film of a given thickness. As we approach the radius, it is very obvious that point in the mold is subjected to greater heat and becomes a hot spot. At that hot spot, the difficulty is in producing a film that has the required strength, because of its very small thickness, to withstand any possible gaseous pressure coming from the mold side. It is distinctly possible in some cases where the angle is very acute that the film is not formed at all, or is re-dissolved during pouring. Probably when the two sides of the angle are solidified, there is expansion in the sand taking place which is tending to open the radius, and if one can imagine an extremely thin film which cannot get thicker because the sand is too hot at that point and a tendency for the face of the mold on the other hand to expand, then the likelihood exists that mold gases may penetrate such film. I call this a blow. A blow, as I see it, on the radius is not in the same family as the type of porosity for which Mr. Sims is endeavoring to give us a fundamental hypothesis. So I would rather leave such a case to further thought and not try to mix things which probably are not mixable.

I feel very grateful to Mr. Sims because once again he has repeated just exactly what he has done every year, he has handed us a fundamental reason for some trouble which will stand inspection from a practical viewpoint. We may not altogether agree with him, but I am perfectly sure if we will take the lesson which is there, that is, insist on adequate boiling of our steels to degasify them, and avoid anything all the way through the whole practical process which increases the possibility of gas absorption, we are bound to receive benefit from this particular type of reasoning.

E. C. TROY[6]: Like everyone else, I can only say that the theory fits and fits well. I came up here to thank the authors and to offer at least one experience that might be a part answer to Mr. Caine's suggestions.

Occasionally we make 0.15 per cent carbon steel carrying about 2.00 to 2.25 per cent silicon. On one occasion, the melter, feeling secure because of the large amount of silicon, made his heat without a boil. He melted "flat" added the 2.00 per cent silicon as ferro-silicon and tapped, adding 0.15 per cent of aluminum to the metal in the ladle. Because of the large volume of dissolved gas, the metal backed out of the mold in a rimming action and pin-holes were found all over the surface of the casting. The chemical laboratory spoon test, cast in a chill mold, was solid. This experience would, it seems to me, check the mechanism described in this paper. Would Mr. Sims tell us why such deoxidizers as calcium do not protect the green sand castings against pin-hole formations?

MR. SIMS: I hesitate to say very much about calcium at the present time because my knowledge is not as complete as I would like

[6] Metallurgist, Dodge Steel Co., Philadelphia, Pa.

but my observations have led me to believe, and I have very good reasons, that calcium is not a deoxidizer for steel. One can secure beneficial effects from calcium which make its use worth while but calcium is not a deoxidizer. I actually poured rimming steel treated with 0.15 per cent calcium and it rimmed better than the other half of the heat which was not treated.

MR. JUPPENLATZ: At the present time aluminum is a rather scarce article. Do you know of any other substitute that the foundry could use to prevent pin-hole formation probably due to the hydrogen of the molds?

MR. SIMS: I have inquired, in regard to the scarcity of aluminum, and the information I received indicates that there will be no trouble getting aluminum for deoxidation. Secondary aluminum is practically as good as primary aluminum or pure aluminum. The impurities that you are likely to encounter in any secondary aluminum are of such nature that they probably have no important effect on steel. The small quantities of iron, silicon, copper and manganese, which constitute the usual impurities, are not likely to cause any trouble. Zinc and magnesium are sometimes present but these elements apparently boil off quickly with no after effects. Zirconium and titanium are somewhat stronger than silicon as deoxidizers but just how much stronger I cannot say. They are not as strong as aluminum and cost more. I think that aluminum is still going to be available.

Classification of Foundry Cost Factors

Report of A.F.A. Cost Committee

At the meeting of the A.F.A. Cost Committee held in 1935, the committee decided that its immediate future activity should be that of comparing the cost classification items of the four major foundry branches, namely, steel, malleable, cast iron and nonferrous, showing in simple form the basic fundamentals of the four branches and the basic similarities between them. The previous work of the committee has been outlined in the report presented before the convention held at Toronto[1].

Since that time the committee has presented for discussion, at the annual conventions, sections of the present report, these covering melting, molding and core cost classifications. Revisions have been made in these classifications and the Committee herewith submits its complete report (Appendix A) for discussion by the membership of the Association.

A review of some points in the earlier history of A.F.A. cost work should, we believe, prove enlightening and, accordingly, we are giving this material here by quoting from the 1935 report:

> "The subject of foundry cost methods was one of the first given consideration at an A.F.A. convention after its organization in 1896 and this discussion resulted in the formation of an A.F.A. Cost Committee which, at the Philadelphia Convention in 1907, presented its first report recommending the preparation of a chart outlining the main divisions of costs. At this same meeting, Ellsworth M. Taylor presented a paper on 'Uniform Foundry Costs.' From that time on, practically every convention has had papers or committee reports on this subject, mainly devoted to fostering uniform cost systems.
>
> "In 1914, a Cost Committee was appointed with A. O. Backert as chairman. The committee's objective was to com-

[1] *Report of A.F.A. Cost Committee*, TRANSACTIONS, American Foundrymen's Association, vol. 43, pp. 584-590 (1935).

NOTE: This report was presented at a session on Foundry Costs, at the 45th Annual A.F.A. Convention, New York City, N. Y., May 12, 1941.

pile a report that would present a cost system 'which would be simple, yet sufficiently comprehensive for satisfactory application to cost accounting in almost every casting plant.' This committee reported at the 1915 convention[2].

"The 1915 report of the Cost Committee resulted in the employing of a cost expert, C. E. Knoeppel & Co., and the preparation of a cost system, which at first was available to those foundries contributing to a special fund. Knoeppel & Company prepared an elaborate cost system and serviced the foundries which had contributed to the special fund. However, in 1919, this cost system was given to the general membership of the A.F.A. and was printed in the *Transactions*[3].

"It was soon realized that this system was not perfect, especially for the smaller foundries, and the Cost Committee continued its efforts to throw more light on systems which were better adapted for the smaller foundries. At the June, 1922, convention, the committee presented an outline form for a simple system[4].

"In the succeeding years the Cost Committee realized that no one system would be suitable for all classes of foundries and it enunciated certain basic principles[5]. Later meetings other papers were presented and discussed[6,7,8,9,10,11,12,13,14].

---

[2] *Report of A.F.A. Committee on Uniform Costs*, TRANSACTIONS, American Foundrymen's Association, vol. 24, pp. 61-73 (1915).

[3] *"Cost Accounting System,"* TRANSACTIONS, American Foundrymen's Association, vol. 28, pp. 65-140 (1919).

[4] *"Report of Cost Committee,"* TRANSACTIONS, American Foundrymen's Association, vol. 30, pp. 282-304 (1922).

[5] *"Report of Cost Committee,"* TRANSACTIONS, American Foundrymen's Association, vol. 32, part 1, pp. 29-31 (1925).

[6] May, H. B., *"Departmental Costs in the Foundry,"* TRANSACTIONS, American Foundrymen's Association, vol. 32, p. 1, pp. 32-43.

[7] Corbett, W. J., *"Cost Finding in a Foundry,"* TRANSACTIONS, American Foundrymen's Association, vol. 32, part 1, pp. 44-78.

[8] Runge, E. T., *"An Explanation of a Foundry Cost Method,"* TRANSACTIONS, American Foundrymen's Association, vol. 33, pp. 148-154 (1925).

[9] McCullough, E. W., *"Progress in Cost Accounting in Industry,"* TRANSACTIONS, American Foundrymen's Association, vol. 34, pp. 9-24 (1926).

[10] Belt, R. E., *"Cost Systems for a Small Foundry,"* TRANSACTIONS, American Foundrymen's Association, vol. 34, pp. 25-44 (1926).

[11] Carter, J. L., *"A Uniform Cost System for a Local Group,"* TRANSACTIONS, American Foundrymen's Association, vol. 35, pp. 20-31 (1927).

[12] Wessling, A., *"Seven Years of Foundry Cost Accounting,"* TRANSACTIONS, American Foundrymen's Association, vol. 35, pp. 32-37.

[13] McDaniel, Don, *"Group Foundry Costs,"* TRANSACTIONS, American Foundrymen's Association, vol. 35, pp. 38-44 (1927).

[14] *"Cost Finding Practice for Steel Foundries,"* (System prepared for Steel Founders' Society), TRANSACTIONS, American Foundrymen's Association, vol. 38, pp. 577-628 (1930).

"Later it was decided that the A.F.A. Cost Committee should do all in its power to foster the use of the cost systems which had been developed by the Steel Founders' Society, the Malleable Iron Research Institute, and the recently formed Gray Iron Institute.

"With the forming of the Gray Iron Institute, the three ferrous branches of the casting industry were provided for and the A.F.A. Committee was put under the chairmanship of J. L. Wick, Jr., and meetings held to interest the non-ferrous foundrymen in adopting a uniform cost system. At the cost session held at the 1930 convention, C. S. Humphreys presented a talk on 'Organizing a Non-Ferrous Foundry Cost Group.' At this same meeting, the Standard Cost System for the recently formed Gray Iron Founders' Society was presented and discussed[15].

"A joint cost conference was held at the 1932 convention with the Gray Iron Institute, at which A. E. Grover presented an extensive paper[16] comparing cost methods.

"At the 1933 convention the Cost Committee, under the chairmanship of J. L. Wick, Jr., presented a play, the purpose of which was to stress the importance of true cost data."

It was the consensus of the 1935 meeting that the committee should continue to promote the use of uniform cost methods and the present report has been developed with that purpose in mind.

The membership of the committee at present consists of—

R. L. Lee, Liberty Foundry Div., Grede Foundries, Inc., Wauwatosa, Wis., *Chairman* (Representing Gray Iron Founders' Society)

J. D. Burlie, Western Electric Co., Chicago, Ill. (Representing Gray Iron Division)

R. L. Collier, Steel Founders' Society of America, Cleveland, O. (Representing Steel Founders' Society)

W. J. Corbett, Atlas Steel Casting Co., Buffalo, N. Y. (Representing Steel Founders' Society)

---

[15] TRANSACTIONS, American Foundrymen's Association, vol. 38, pp. 577-628 (1930).

[16] "*Joint Conference on Costs*," TRANSACTIONS, American Foundrymen's Association, vol. 40, pp. 461-482 (1932).

V. L. Diefenbacher, Hamilton Foundry & Machine Co., Hamilton, O. (Representing Gray Iron Founders' Society and Gray Iron Division)

G. E. Hels, Vilter Manufacturing Co., 2217 S. First St., Milwaukee, Wis. (Representing Gray Iron Division)

S. Kitto, Belle City Malleable Iron Co., Racine, Wis. (Representing Malleable Founders' Society and Malleable Division)

E. J. Metzger, Falcon Bronze Co., Youngstown, O. (Representing Non-Ferrous Division)

C. S. Roberts, Dodge Steel Co., Philadelphia, Pa. (Representing Steel Founders' Society)

E. H. Roeming, Chain Belt Co., Gray Iron Castings Div., Milwaukee, Wis. (Representing Gray Iron Division)

J. A. Wagner, Wagner Malleable Iron Co., Decatur, Ill. (Representing Malleable Division)

Respectfully submitted,

RALPH L. LEE, *Chairman*

Appendix A — 1941 A.F.
Comparison of Foundry Cost Classifications for Gr

MELTING COST

Gray Iron	Malleable
Metals	**Metals**
Pig Iron	Pig Iron
Purchased Scrap	Purchased Scrap
Own Scrap	Own Scrap
Returns	Ferromanganese
Ferromanganese	Ferrosilicon
Ferrosilicon	
Special alloys are charged to individual jobs that contain them.	Special alloys are charged to individual jobs that contain them.
The cost includes freight, cartage, unloading, and storage labor.	The cost includes freight, cartage, unloading, and storage labor.
Fuels	**Fuels**
Coke, electricity, coal, oil, etc.	Coke, electricity, coal, fuel, oil, gas, etc.
Embracing the cost of all melting fuels, including freight, cartage, unloading and storage labor.	Embracing the cost of all melting fuels, including freight, cartage, unloading and storage labor.
In case of electricity, include the demand charge.	In case of electricity, include the demand charge and secondary charge of current consumed.
Refractories for Melting Equipment and Ladles	**Refractories for Melting Equipment and Ladles**
Fire Clay	Fire Clay
Firebrick	Firebrick
Fire Stone	
Ganister	
Fluxes	**Fluxes**
Lime—Limestone	Lime—Limestone
Fluorspar	Iron ore
Oyster or clam shells	
Special trade name fluxes	Special trade name fluxes

Cost Committee Report
Iron, Malleable, Steel and Non-Ferrous Foundries

MELTING COST

Steel	Non-Ferrous
Metals	**Metals**
Pig Iron	Virgin Metal
Purchased Scrap	Purchased Ingots
Own Scrap	Purchased Scrap
Ferromanganese	Own Scrap
Ferrosilicon	Returns
Special alloys are charged to individual jobs that contain them.	Special alloys are charged to individual jobs that contain them.
The cost includes freight, cartage, unloading, and storage labor.	The cost includes freight, cartage, unloading, and storage labor.
Fuels	**Fuels**
Fuel oil for furnaces and ladles	Coke, electricity, coal, oil, gas, etc.
Gas for furnaces and ladles	
Electricity, etc.	
Embracing the cost of all melting fuels, including freight, cartage, unloading and storage labor.	Embracing the cost of all melting fuels, including freight, cartage, unloading and storage labor.
In case of electricity, include the demand charge.	In case of electricity, include the demand charge.
Refractories for Melting Equipment and Ladles	**Refractories for Melting Equipment and Ladles**
Fire Clay	Fire Clay
Firebrick	Firebrick
Fire Stone	Fire Stone
Ganister	Ganister
Ladle Brick	Ladle Sleeves
Ladle Loam	Stoppers
Sleeves	Nozzles
Stoppers	Crucibles
Nozzles	High temperature refractories
	Insulating materials
Fluxes	**Fluxes**
Lime—Limestone	Glass
Iron ore	Oxides
Oyster or clam shells	Carbon
Aluminum for deoxidizing	
Fluorspar	
Special trade name fluxes	Special trade name fluxes

MELTING COST—(Cont'd)

Gray Iron

Misc. Supplies and Tools
Embracing the cost of all misc. supplies and tools used at the melting furnace or cupola.

Laboratory Expense
All work performed by laboratory for Melting Department such as determinations, testing analysis, photomicrographs, etc. If laboratory work is purchased outside, Melted Metal Department is charged for purchases applicable to department.

Labor
Includes all labor in melting dept. from delivery of raw materials to the melted iron at the spout, such as transporting materials to charging floor, hauling out slag and ashes, patching cupola, cupola tending, yard labor, charging, relining cupola, cranemen.

Not including delivery of metal from spout or mixing ladle.

Malleable

Misc. Supplies and Tools
Embracing the cost of all misc. supplies and tools used at the melting furnace or cupola.

Laboratory Expense
All work performed by laboratory for Melting Dept. such as determinations, testing analysis, photomicrographs, etc. If laboratory work is purchased outside, Melted Metal Dept. is charged for purchases applicable to department.

Labor
To include all labor identified with the melting of metal.

Accounts Comm All

The following accounts are common to all departments and are explained in detail **su**

Gray Iron	Malleable
Supervision	Supervision
Clerks	Clerks
Repairs to buildings	Repairs to buildings
Repairs to equipment and machinery	Repairs to equipment and machinery
Compensation insurance	Compensation insurance
Payroll taxes	Payroll taxes
Vacation expense	Vacation expense
Power—Heat—Light	Power—Heat—Light
Plant general	Plant general
Maintenance	Maintenance
Total fixed charges	Total fixed charges
Administrative and selling	Administrative and selling

MELTING COST—(Cont'd)

Steel

Misc. Supplies and Tools
Embracing the cost of all misc. supplies and tools used at the melting furnace or cupola.

Laboratory Expense
All work performed by laboratory for Melting Dept. such as determinations, testing analysis, photomicrographs, etc. If laboratory work is purchased outside, Melted Metal Dept. is charged for purchases applicable to department.

Labor
To include all labor identified with the melting of metal.

Non-Ferrous

Misc. Supplies and Tools
Embracing the cost of all misc. supplies and tools used at the melting furnace or cupola.

Laboratory Expense
All work performed by laboratory for Melting Dept. such as determinations, testing analysis, photomicrographs, etc. If laboratory work is purchased outside, Melted Metal Dept. is charged for purchases applicable to department.

Labor
To include all labor identified with the melting of metal.

All Departments

supplementary sheets:

Steel

Supervision
Clerks
Repairs to buildings
Repairs to equipment and machinery
Compensation insurance
Payroll taxes
Vacation expense
Power—Heat—Light
Works Burden (Plant general)
Maintenance
Total fixed charges
Administrative and selling

Non-Ferrous

Supervision
Clerks
Repairs to buildings
Repairs to equipment and machinery
Compensation insurance
Payroll taxes
Vacation expense
Power—Heat—Light
Plant general
Maintenance
Total fixed charges
Administrative and selling

Yield

Gray Iron

Yield is the per cent of shippable castings in pounds to the molten metal in pounds used to produce these castings.

Molten metal is the metal charged less all metal weight unaccounted for due to dirty scrap (*unrecovered spillage, grinding and chipping of gates and risers).

* While unrecovered spillage, grinding and chipping of gates and risers is a loss that occurs after the metal is melted, as a matter of convenience, this loss is considered melting loss. This loss runs from 3 to 8 per cent.

Yield equals shippable castings in lb.

Divided by

Shippable Castings in Pounds { plus } Weight of Defective castings and weight of gates, risers, etc. of shippable and defective castings in pounds.

Malleable

Yield of a given job is the percentage relation of the weight of good production to the weight of metal used to produce the job. Yield of any given job, in percentage of metal charged, is obtained by dividing the pounds of good production by the pounds of metal poured, increased by the average oxidation and metal loss, including finishing loss.

Example—18000-lb. metal poured
9.4 per cent metal loss
9000-lb. good finished castings produced

$18000 \div 90.6 \ (100-9.4) = 19868$
$9000 \div 19868 = 45$ per cent yield

Distribution of Costs of Metal and Melting

The variable portion of the cost of metal (*i.e.* the oxidation and metal Loss) and all costs of melting are distributed to jobs on the basis of the weight of metal charged for each.

The non-variable portion of the cost of metal (*i.e.*, all costs other than oxidation and metal loss) is distributed to jobs on the basis of the weight of good production.

Yield

Steel

Yield means the percentage of total metal charged into the furnace that results in good castings. To obtain the yield of a specific casting, the following data must be known: Shipping weight of the casting, the weight of the heads and gates used on the casting and the weight of the defective castings, including the weights of the heads and gates on them.

Yield for the casting then can be calculated by using the following formula:

$$\frac{G \times (100-(S+L+B)}{G+D+H} = Per\ Cent\ Yield$$

G = Shipping weight of casting

*S = Normal metal losses

*L = All metal unaccounted for

*B = Slagged castings

D = Weight of defective castings in pounds

H = Weight of heads-gates used on both good and defective castings in pounds

*These are in percentage of total metal charged.

Non-Ferrous

To find the yield of good castings resulting from any amount of metal melted, divide the total weight of finished castings by the total weight charged into the furnace.

Factors which would cause different yields are the proportion of gates, sprues, and risers to the casting proper, the length of time the melt is held in the furnace, the temperature used, the quality of the metal, and the alloy, and the filing, grinding and finishing to be done.

A simple formula would be:

Yield equals

Weight of castings in finished form.

Divided by

Total weight charged into Furnace.

MOLDING COST

Gray Iron	Malleable
Direct Labor	**Direct Labor**
To include all:	To include all:
Molders, dry and green sand	Molders, dry and green sand
Molders helpers	Molders helpers
Apprentices	Apprentices
Finishers	Finishers
Rammers	Rammers
Slinger operators and helpers	Slinger operators and helpers
Core setters	Core setters
Pouring, whether done by molders, helpers, or pouring gang	Pouring, whether done by molders, helpers, or pouring gang
	Lifting and carrying out molds
Indirect Labor	**Indirect Labor**
All other labor in the molding department, including:	All other labor in the molding department, including:
Cranemen	Cranemen and truckmen
Shakeout labor	
Moving flasks, jackets, weights and boards	Moving flasks, jackets, weights and boards
Breaking or cutting gates from castings	Breaking or cutting gates from castings
Cutting and tempering sand, testing, etc.	Cutting and tempering sand, testing, etc.
Facing sand mixers	Facing sand mixers
Weight shifters and dumpers	Weight shifters and dumpers
Mold or drying oven operators and tenders	
Wheeling and conveying sand, sprues and castings	Wheeling and conveying sand sprues and castings
Tool room tenders and labor	
Gaggermen	
Sand mill operators	
Fitting patterns to flasks	Fitting patterns to flasks
Mold checkers and inspectors	
Dry floor laborers	
Shankpushers	
	Relining ladles
	Making and repairing flasks, jackets and bottom boards
	Cleaning floors and gangways
	Making matches
Other misc. foundry labor not charged to direct labor	Other misc. foundry labor not charged to direct labor

MOLDING COST

Steel	Non-Ferrous
Direct Labor	**Direct Labor**
To include all:	To include all:
Molders, dry and green sand	Molders, dry and green sand
Molders helpers	Molders helpers
Apprentices	Apprentices
Finishers	Finishers
Rammers	Rammers
Slinger operators and helpers	Slinger operators and helpers
Core setters	Core setters
	Pouring if done by molders or helpers
Mold clampers	
	Lifting and carrying out molds
Indirect Labor	**Indirect Labor**
All other labor in the molding department, including:	All other labor in the molding department, including:
Pouring labor	Pouring, if done by separate pouring gangs
Cranemen	Cranemen and monorail operators
Shakeout labor	Shakeout labor
Moving flasks, jackets, weights and boards	Moving flasks, jackets, weights and boards
	Breaking or cutting gates from castings
Cutting and tempering sand, testing, etc.	Cutting and tempering sand, testing, etc.
Facing sand mixers	Facing sand mixers
	Weight shifters and dumpers
Mold or drying oven operators and tenders	Mold or drying oven operators and tenders
Wheeling and conveying sand, sprues and castings	Wheeling and conveying sand, sprues and castings
Tool room tenders and labor	Tool room tenders and labor
Gaggermen	
Sand mill operators	
Fitting patterns to flasks	Fitting patterns to flasks
Mold checkers and inspectors	
Dry floor laborers	
Heat followers	
Molding general labor	
Molding machine repairmen	
Runner cup makers	
Other misc. foundry labor not charged to direct labor	Other foundry labor not charged to direct labor

MOLDING COST—(Cont'd)

Gray Iron

Molding Sand

Cost of molding sand, including freight, cartage and unloading

Misc. Supplies and Tools

Air hose and fittings
Bar iron and steel
Fireclay and bentonite
Chaplets, chills and clamps
Chains, bolts and gaggers
Lubricants and belt dressing
Nails and lumber
Shovels, riddles and brushes
Rammers and rammer butts
Other compounds used in molding sand mixtures
Silica flour, plumbago, sea coal
Incandescent lamps
Fuel for dry sand ovens
Fuel to dry pit molds
Waste, tools, etc.

Flasks

Labor and material used for making, altering or repairing wood or iron flasks, bottom boards, plates, squeeze boards, bands, jackets, etc., which are not charged direct to customer. Whenever possible the cost of special equipment should be charged direct to customer.

Patterns and Pattern Dept.

All patternmaking labor and materials for making, altering, or repairing patterns, mounting on boards, plates, etc., which are not chargeable to a customer.

Molding department charged on basis of patterns hours applied against molding, overhead included, to cover cost of patterns made or repaired for molding department, such as weight patterns, clamp patterns, etc.

Malleable

Molding Sand

Cost of molding sand, including freight, cartage and unloading

Misc. Supplies and Tools

Air hose and fittings
Bar iron and steel
Fireclay and bentonite
Chaplets, chills and clamps
Chains, bolts and gaggers
Lubricants and belt dressing
Nails and lumber
Shovels, riddles and brushes
Rammers and rammer butts
Other compounds used in molding sand mixtures
Silica flour, flour
Incandescent lamps

Waste, tools, etc.

Flasks

Labor and material used for making new wood flasks and repairing them, including bottom boards, plates and cross bars.

Preferably a reserve fund should be created for flasks by charging to costs monthly an amount based upon a predetermined rate per net ton of castings produced.

The actual expenditure for new flasks and repairs should be charged to this reserve fund. The cost of metal flasks is generally charged to a capital account.

Patterns and Pattern Dept.

Pattern lumber and supplies when not chargeable to a customer. Making, altering or repairing patterns when not chargeable to a customer.

MOLDING COST—(Cont'd)

Steel — Other Indirect Expenses

Molding Sand

Cost of molding sand, including freight, cartage and unloading

Misc. Supplies and Tools

Air hose and fittings
Bar iron and steel
Fireclay and bentonite
Chaplets, chills and clamps
Chains, bolts and gaggers
Lubricants and belt dressing
Nails and lumber
Shovels, riddles and brushes
Rammers and rammer butts
Other compounds used in molding sand mixtures
Silica flour, flour
Incandescent lamps

Waste, tools, etc.

Flasks

Labor and material used for making new wood flasks and repairing them, including bottom boards, plates and cross bars.

Preferably a reserve fund should be created for flasks by charging to costs monthly on amount based upon a predetermined rate per net ton of castings produced. The actual expenditure for new flasks and repairs should be charged to this reserve fund. The cost of metal flasks is generally charged to a capital account.

Patterns and Pattern Dept.

Cost of repairing, rigging and mounting patterns, including the making of heads and gates, when such expense is not chargeable to a customer.

Pattern expense not chargeable to customer is charged to molding department, core department, and special charges, in proportion to work performed for departments, balance is charged to Works Burden.

Non-Ferrous

Molding Sand

Cost of molding sand, including freight, cartage and unloading

Misc. Supplies and Tools

Air hose and fittings
Bar iron and steel
Fireclay and bentonite
Chaplets, chills and clamps
Chains, bolts and gaggers
Lubricants and belt dressing
Nails and lumber
Shovels, riddles and brushes
Rammers and rammer butts
Other compounds used in molding sand mixtures
Silica flour, flour
Incandescent lamps

Waste, tools, etc.

Flasks

All flask repairs and alterations, including material and labor not chargeable to individual job cost.

Patterns and Pattern Dept.

All patternmaking labor and materials for making, altering, or repairing patterns, mounting on boards, plates, etc., which are not chargeable direct to a customer.

Whenever possible the cost of equipment should be charged direct to the customer.

Pattern expense not chargeable to customer is charged to molding department, core department, and general overhead expenses.

MOLDING COST—(Cont'd)

Accounts Co

The following accounts are common to all departments and are explained in det

Gray Iron	Malleable
Supervision	Supervision
Clerks	Clerks
Repairs to Buildings	Repairs to Buildings
Repairs to Equipment and Machinery	Repairs to Equipment and Machinery
Compensation Insurance	Compensation Insurance
Payroll Taxes	Payroll Taxes
Vacation Expense	Vacation Expense
Power—Heat—Light	Power—Heat—Light
Plant General	Plant General
Maintenance	Maintenance
Total Fixed Charges	Total Fixed Charges
Administrative and Selling	Administrative and Selling

Distribution of Indirect Costs in Molding Department

Gray Iron

The Molding Department overhead rates are established as a percentage of the molding direct labor cost.

Malleable

Those items which generally follow molding direct labor (such as parting, chaplets, shifting, cutting and tempering sand) are distributed on that basis; those items which generally follow weight of metal poured (such as molding sand, bonding clay, delivering metal, handling sprue, relining ladles) are distributed on that basis, and those items which largely follow the number of jobs started (such as fitting patterns to flasks, flask and pattern handling) are distributed on that basis.

MOLDING COST—(Cont'd)

Departments

upplementary sheets:

Steel	Non-Ferrous
Supervision	Supervision
Clerks	Clerks
Repairs to Buildings	Repairs to Buildings
Repairs to Equipment and Machinery	Repairs to Equipment and Machinery
Compensation Insurance	Compensation Insurance
Payroll Taxes	Payroll Taxes
Vacation Expense	Vacation Expense
Power—Heat—Light	Power—Heat—Light
Works Burden (Plant General)	Plant General
Maintenance	Maintenance
Total Fixed Charges	Total Fixed Charges
Administrative and Selling	Administrative and Selling

Distribution of Indirect Costs in Molding Department

Steel	Non-Ferrous
The Molding Department overhead rate is established as a percentage of the molding direct labor cost.	The Molding Department overhead rate is established as a percentage of the molding direct labor cost.

CORE COST

Gray Iron	Malleable
Direct Labor	**Direct Labor**
Core makers	Core makers
Core pasters	Core pasters and painters
Core assemblers	Core assemblers
Core gaugers	
Indirect Labor	**Indirect Labor**
Sand handling and preparation	Sand handling and preparation
Core Carriers	Core carriers
Core inspectors	Core inspectors
Core oven tenders	Core oven tenders
Core box tenders	
Cranemen	
Plate handlers	
Cleaning cores	Cleaning cores
Wire cutters and straighteners	Wire cutters and straighteners
	Sweepers
	Clean up and disposal labor
General labor	
Other labor not charged to direct labor	
Core Sand	**Core Sand**
Cost of core sand, including freight, cartage, and unloading	Cost of core sand, including freight, cartage, and unloading
Core Fuel	**Core Fuel**
Cost of fuel for core ovens including freight, cartage and unloading	Cost of fuel for core ovens, including freight, cartage and unloading
Misc. Supplies and Tools	**Misc. Supplies and Tools**
Core oil and compounds	Core oil and compounds
Shovels, riddles and brushes	Shovels, riddles and brushes
Core wire and rods	Core wire and rods
Core plates and nails	Core plates and nails
Glutrin and molasses	Rosin and pitch
Rammers and rammer butts	
	Bonding material
Tools, etc.	Tools, etc.

CORE COST

Steel	Non-Ferrous
Direct Labor	**Direct Labor**
Core makers	Core makers
Core pasters	Core pasters
Core assemblers	Core assemblers
Apprentices	
Indirect Labor	**Indirect Labor**
Sand handling and preparation	Sand handling and preparation
Core carriers	Core carriers
Core inspectors	Core inspectors
Core oven tenders	Core oven tenders
Core box tenders	Core box tenders
Cranemen	Cranemen
Core gaugers	Core gaugers
Plate handlers	Plate handlers
Cleaning cores	Cleaning cores
Wire cutters and straighteners	Wire cutters and straighteners
Chainmen	
General labor	General labor
Other labor not charged to direct labor	Other labor not charged to direct labor
Core Sand	**Core Sand**
Cost of core sand, including freight, cartage, and unloading	Cost of core sand, including freight, cartage, and unloading
Core Fuel	**Core Fuel**
Cost of fuel for core ovens, including freight, cartage and unloading	Cost of fuel for core ovens, including freight, cartage and unloading
Misc. Supplies and Tools	**Misc. Supplies and Tools**
Core oil and compounds	Core oil and compounds
Shovels, riddles and brushes	Shovels, riddles and brushes
Core wire and rods	Core wire and rods
Core plates and nails	Core plates and nails
Molasses	Molasses
Rammers and rammer butts	Rammers and rammer butts
Air hose and fittings	
Bar iron and steel	
Chains and bolts	
Silica flour, flour	
Chills—Fireclay	
Incandescent lamps	
Lubricants—Lumber	
Tools, etc.	Tools, etc.

CORE COST—(Cont'd)

Gray Iron

Patterns and Pattern Dept.

All core box making labor and materials for making, altering or repairing boxes, which are not chargeable to a customer.

Core department charged on basis of pattern hours applied against core making, overhead included, to cover cost of patterns made or repaired for core dept.

Malleable

Patterns and Pattern Dept.

Making, altering, or repairing core boxes when not chargeable to a customer.

Accounts Co

The following accounts are common to all departments and are explained in det

Gray Iron	Malleable
Supervision	Supervision
Clerks	Clerks
Repairs to Buildings	Repairs to Buildings
Repairs to Equipment and Machinery	Repairs to Equipment and Machinery
Compensation Insurance	Compensation Insurance
Payroll Taxes	Payroll Taxes
Vacation Expense	Vacation Expense
Power—Heat—Light	Power—Heat—Light
Plant General	Plant General
Maintenance	Maintenance
Total Fixed Charges	Total Fixed Charges
Administrative and Selling	Administrative and Selling

Distribution of Ind

The core department overhead rate is established as a percentage of the core direct labor cost.

Cost of producing core material should be omitted from overhead and computed on a weight basis and charged directly to each job on this basis.

Those items which generally follow core direct labor (such as pasting and painting, inspection and counting, wire cutters and straighteners) are distributed on that basis, and those items which generally follow weight of cores used (such as core sand, core fuel, sand handling and preparation, oven tenders) are distributed on that basis.

CORE COST—(Cont'd)

Steel	Non-Ferrous
Patterns and Pattern Dept.	Patterns and Pattern Dept.
Making, altering, or repairing core boxes when not chargeable to a customer.	Making, altering, or repairing core boxes when not chargeable to a customer.

Departments

pplementary sheets:

Steel	Non-Ferrous
Supervision	Supervision
Clerks	Clerks
Repairs to Buildings	Repairs to Buildings
Repairs to Equipment and Machinery	Repairs to Equipment and Machinery
Compensation Insurance	Compensation Insurance
Payroll Taxes	Payroll Taxes
Vacation Expense	Vacation Expense
Power—Heat—Light	Power—Heat—Light
Works Burden (Plant General)	Plant General
Maintenance	Maintenance
Total Fixed Charges	Total Fixed Charges
Administrative and Selling	Administrative and Selling

in Core Department

The core department overhead rate is established as a percentage of the core direct labor cost	The core department overhead rate is established as a percentage of the core direct labor cost

CLEANING AND FINISHING COST

Gray Iron

The cleaning and finishing department is sub-divided into the following departments, each of which bears the proper proportion of labor (both direct and indirect), supplies and tools, etc.

Tumbling
Hand Blast
Automatic Blast
Tumble Blast
Table Blast
Grinding
Pickling
Finishing and Inspecting
Chipping-air-hand
Shipping

Malleable

The cleaning and finishing department is sub-divided into the following departments, each of which bears the proper proportion of labor (both direct and indirect), supplies and tools, etc.

Hard Iron Cleaning
Hard Iron Trimming
Annealing
Soft Iron Cleaning
Grinding
Finishing
Assorting and Shipping

Accounts Co

The following accounts are common to all departments and are explained in det

Gray Iron

Supervision
Clerks
Repairs to Buildings
Repairs to Equipment and Machinery
Compensation Insurance
Payroll Taxes
Vacation Expense
Power—Heat—Light
Plant General
Maintenance
Total Fixed Charges
Administrative and Selling

Malleable

Supervision
Clerks
Repairs to Buildings
Repairs to Equipment and Machinery
Compensation Insurance
Payroll Taxes
Vacation Expense
Power—Heat—Light
Plant General
Maintenance
Total Fixed Charges
Administrative and Selling

CLEANING AND FINISHING COST

Steel	Non-Ferrous
Direct Labor	**Direct Labor**
To include all:	To include all:
Stand Grinders	Cutters
Swing Grinders	Tumblers
Chippers	Blasters
Machine Operators	Grinders
Cutting Operators	Filers, etc.
Pressman and helpers	
Indirect Labor	**Indirect Labor**
All other labor in this department, including:	All other labor in this department
Blast helpers	
Chainmen	
Cranemen	
Electric Welders	
Power truck operators, etc.	
Misc. Supplies and Tools	**Misc. Supplies and Tools**
Acetylene and carbide	Sand blast sand
Air hose and fittings	Grinding wheels
Blasting abrasives, hose and nozzles	Saw, files, acids, etc.
Brushes and chains	
Chisels and cutters	
Grinding wheels, hammers, sledges, handles, lamps, lubricants, oxygen, tools, etc.	

Departments

supplementary sheets:

Steel	Non-Ferrous
Supervision	Supervision
Clerks	Clerks
Repairs to Buildings	Repairs to Buildings
Repairs to Equipment and Machinery	Repairs to Equipment and Machinery
Compensation Insurance	Compensation Insurance
Payroll Taxes	Payroll Taxes
Vacation Expense	Vacation Expense
Power—Heat—Light	Power—Heat—Light
Works Burden (Plant General)	Plant General
Maintenance	Maintenance
Total Fixed Charges	Total Fixed Charges
Administrative and Selling	Administrative and Selling

MISCELLANEOUS DEPARTMENTS

Gray Iron	Malleable
Heat Treating	**Heat Treating**
Labor	Shown under cleaning and finishing
Fuel	
Supplies and Tools	
Pattern	**Pattern**
Direct Labor	Patterns (not chargeable directly to customers). To include the cost of all patterns and pattern work, the total expense of which should be charged directly to the job.
Direct Material	
Supplies and Tools	
Supervision	
Clerks	

MISCELLANEOUS DEPARTMENTS

Steel

Heat Treating
Labor
Fuel
Supplies and Tools

Pattern
Direct Labor
Direct Materials
Supplies and Tools
Supervision
Clerks

Non-Ferrous

Heat Treating
Labor
Fuel
Supplies and Tools

Pattern
Direct Labor
Direct Materials
Supplies and Tools
Supervision
Clerks

SUPPLEMENTARY SHEETS

Showing detail of accounts common to all departments

Gray Iron

Supervision—Clerks
Foremen and assistant foremen in department.

Department clerks.

Repairs to Buildings
Repairs to buildings usually charged to general overhead expense.

Repairs to Equipment and Machinery
Labor, material, and other charges for repairing all equipment in Department, such as rewinding motors, repair parts, etc.

Compensation Insurance
Allocated in the ratio of the payroll of the Department to total plant payroll.

This charge also includes occupational disease insurance which is included under compensation insurance in some states, while in other states it is a separate charge or tax.

Payroll Taxes
Old age benefit tax

Unemployment insurance

Allocated on the same basis as compensation insurance

Vacation Expense
To include the vacation expense of all productive employes in this department

Power—Heat—Light
Power—Allocated on basis of consumption

Heat—Allocated in the ratio of heated area of the department to total heated area

Light—Allocated on basis of consumption

Plant General
To include all forms of misc. plant expense of a general nature not included elsewhere.

Allocated on direct and indirect labor basis, each department absorbs its share

Malleable

Supervision—Clerks
Foremen and assistant foremen in department.

Department clerks.

Repairs to Buildings
Distributed on a floor space basis, each department bears its proportionate share.

Repairs to Equipment and Machinery
Embracing the cost of all labor and materials used in the upkeep and maintenance of equipment.

Compensation Insurance
Allocated in the ratio of the payroll of the Department to total plant payroll.

This charge also includes occupational disease insurance which is included under compensation insurance in some states, while in other states it is a separate charge or tax.

Payroll Taxes
Old age benefit tax

Unemployment insurance

Allocated on the same basis as compensation insurance.

Vacation Expense
To include the vacation expense of all productive employes in this department

Power—Heat—Light
Power—Allocated on basis of consumption

Heat—Allocated in the ratio of heated area of the department to total heated area

Light—Allocated on basis of consumption

Plant General
To include all forms of misc, plant expense of a general nature not included elsewhere.

Allocated to department on basis applicable to the particular expense item

SUPPLEMENTARY SHEETS

Showing detail of accounts common to all departments

Steel	Non-Ferrous
Supervision—Clerks Foremen and assistant foremen in department. Department clerks.	**Supervision—Clerks** Foremen and assistant foremen in department. Department clerks.
Repairs to Buildings Repairs to buildings charged to Works Burden.	**Repairs to Buildings** Carried in fixed overhead. This part of fixed overhead distributed on floor space basis.
Repairs to Equipment and Machinery Labor, material, and other charges for repairing all equipment in department is charged to this department.	**Repairs to Equipment and Machinery** Cost of all labor and materials used in the upkeep and maintenance of all equipment in department.
Compensation Insurance Allocated in the ratio of the payroll of the department to total plant payroll. This charge also includes occupational disease insurance which is included under compensation insurance in some states, while in other states it is a separate charge or tax.	**Compensation Insurance** Allocated in the ratio of the payroll of the department to total plant payroll. This charge also includes occupational disease insurance which is included under compensation insurance in some states, while in other states it is a separate charge or tax.
Payroll Taxes Old age benefit tax Unemployment insurance Allocated on the same basis as compensation insurance	**Payroll Taxes** Old age benefit tax Unemployment insurance Allocated on the same basis as compensation insurance
Vacation Expense	**Vacation Expense** To include the vacation expense of all productive employes in this department
Power—Heat—Light *Power*—Allocated on basis of consumption *Heat*—Allocated on the ratio of heated area of the department to total heated area *Light*—Allocated on basis of consumption	**Power—Heat—Light** *Power*—Allocated on basis of consumption *Heat*—Allocated in the ratio of heated area of the department to total heated area *Light*—Allocated on basis of consumption
Plant General (Works Burden) To include all forms of misc. plant expense of a general nature not included elsewhere. Allocated to department on basis applicable to the particular expense item	**Plant General** To include all forms of misc, plant expense of a general nature not included elsewhere. Allocated to department on basis applicable to the particular expense item

SUPPLEMENTARY SHEETS—(Cont'd)

Showing detail of accounts

Gray Iron	Malleable
Maintenance	**Maintenance**
Each department is charged with maintenance on basis of hours, overhead included, covers minor repairs not included elsewhere	All maintenance is included under the head of "Repairs"
Total Fixed Charges	**Total Fixed Charges**
Depreciation:	Depreciation:
Buildings—Allocated on floor space basis	*Buildings*—Allocated on floor space basis
Equipment—Allocated on valuation basis	*Equipment*—Allocated on valuation basis
Administrative and Selling	**Administrative and Selling**
Total administrative and selling prorated on basis of direct and indirect labor of payroll in each department.	No portion of the general administrative and selling expense is reckoned directly as a department cost, but as these expenses are allocated on total manufacturing costs, each department ultimately bears a share of them.

SUPPLEMENTARY SHEETS—(Cont'd)

mmon to all departments

Steel

Maintenance

Repairs and maintenance are charged to the department where made. General repairs and maintenance are included in the Works Burden

Total Fixed Charges

Depreciation—Fire

Insurance—Real Estate Taxes

Distributed on the basis of land, buildings and capitalized equipment

Administrative and Selling

No portion of the general administrative and selling expense is reckoned directly as a department cost, but as these expenses are allocated on total manufacturing costs, each department ultimately bears a share of them.

Non Ferrous

Maintenance

Repairs and maintenance are charged to the department where made. General repairs and maintenance to general overhead expense

Total Fixed Charges

Taxes, insurance, depreciation and similar items of general plant expense prorated to department on appropriate basis, otherwise carried to general overhead expense

Administrative and Selling

No portion of the general administrative and selling is reckoned directly as a department cost, but as these expenses are allocated on total manufacturing costs, each department ultimately bears a share of them.

A University Course in Foundry Control Methods

FULTON HOLTBY* AND H. F. SCOBIE,* MINNEAPOLIS, MINN.

Abstract

For many years, the University of Minnesota offered a course in foundry practice that was required of students in mechanical engineering. About three years ago, the University made a study of the future of manufacturing processes in the foundry industry which revealed its rapid growth and the need for engineers, laboratory technicians and executives. Two years previously, the University had completely overhauled its required elementary foundry practice course and, after the above mentioned survey, decided to install an elective course for advanced students dealing with foundry control methods. Most of the apparatus required for such a course was built by the University and the course itself included a study of raw materials, processing materials, manufacturing processes, maintenance of control equipment, and a metallurgical inspection of finished cast products. The various points studied, not only in the control course, but in the elementary foundry practice course are outlined herein. In addition, two problem courses are offered to advanced students. These are entitled foundry research problems and advanced engineering problems. These also are elective courses and students in these courses assist in the study of research problems sponsored by the foundry department of the University. The authors outline quite completely the various subjects studied in the different courses and demonstrate the interest in these courses by means of a chart which shows the increase in registration during the past few years in advanced foundry study.

1. Probably no other industry has awakened so rapidly from several thousand years of dormant existence as has the foundry industry. Spurred by outside competition, hard-earned discoveries, and a new generation of technically trained men, the industry has thrown off the shackles of a craft and developed scientific foundry

* Assistant Professor of Mechanical Engineering and Instructor of Foundry Practice, respectively. University of Minnesota.

NOTE: This paper was presented before the Engineering Instructor's Dinner at the 45th Annual A.F.A. Convention, New York City, N. Y., May 12, 1941.

practice. As a result, it has been difficult for the universities and colleges to keep their foundry courses abreast of this recent rapid development of the foundry industry.

OLD TEACHING METHODS DISCARDED

2. Five years ago the University of Minnesota began developing a series of modern foundry practice courses for future engineers. The growing response and continued interest shown by both the students and the industry have been highly encouraging.

3. Prior to 1936, this university had offered only one foundry course, which, in the words of one student, consisted of "pounding sand, pouring book ends, and remelting scrap castings." The course was required of all mechanical engineering students and was taught by one instructor.

4. Today the foundry department offers five major foundry courses, namely, Elementary Foundry Practice, required of all mechanical engineering sophomores; Intermediate Foundry Practice, elective to sophomores; Foundry Control Methods, elective to upper classmen; and two foundry problem courses, Foundry Research Problems and Advanced Engineering Problems, which are elective to upper classmen and graduate students.

5. Now included on the personnel of the foundry department are an assistant professor, an instructor, a laboratory mechanic, three National Youth Administration student-assistants, and seven Works Progress Administration research assistants.

ELEMENTARY FOUNDRY PRACTICE COURSE

6. Each year approximately 225 students enroll in elementary foundry practice, which is required of all mechanical engineering students. The primary purpose of this course is to acquaint all mechanical engineering students with modern foundry practice, the possibilities and limitations of castings for engineering use, and the factors in design of castings.

7. The course consists of 20 hours of lecture and 30 hours of laboratory. Since the University of Minnesota operates on a 10-week quarter system instead of a 15-week semester system, the course is completed in 10 weeks. There are a definite number of operations and projects to be completed, and the course is as carefully laid out as any university course in chemistry or physics. A schedule of assignments is shown in Table 1. Figs. 1 and 2 show students at work on projects. Fig. 3 shows student projects ready for pouring.

FIG. 1—STUDENTS MAKING CORE ROOM PROJECTS.

8. Four points are stressed repeatedly in this elementary course: (1) the development of the foundry industry during the past two decades, with the resulting betterment of products; (2) the importance of control of the foundry processes; (3) the practical knowledge it gives for work in other fields; and (4) the increasing opportunities for employment as a foundry engineer.

9. The idea that this is a new and rapidly developing phase

FIG. 2—STUDENTS OPERATING MOLDING MACHINES.

Table 1

UNIVERSITY OF MINNESOTA 1940 — '41

SCHEDULE FOR ELEMENTARY FOUNDRY PRACTICE

Lecture Number	Lecture Subject	Text[1] Assignment	Laboratory Number	Laboratory Assignment
1	The Casting Industry	1–9	1	Demonstrations and Equipment Inspection.
2	Materials for the Foundry	10–29		
3	Bench Molding Practice	30-39	2	Bench Molding Projects No. 5, 7 and 50.
4	Floor Molding Practice	40–49		
5	Machine Molding Practice	50–67	3	Molding Machine Projects No. 25, 27 and 29.
6	Materials for the Foundry	68–76		
7	Core Practice	77–85	4	Core Room Projects No. 10, 11, 13 and 15.
8	Cupola Melting Practice	88–99		
9	Cupola Melting Practice	100–109	5	Cupola Preparation and Charge Calculation.
10	Mid-quarter Examination	Review		
11	Melting Practice	110–117	6	Cupola Operation and Pouring.
12	Cleaning and Finishing	118–133		
13	Constitution of Cast Iron	135–151	7	Cleaning and Inspection of Castings.
14	Properties of Cast Iron	152–162		
15	Alloy Cast Iron	163–173	8	Electric Furnace Operation and Duplexing.
16	Malleable Iron Castings	192–201		
17	Steel Castings	202–221	9	Non-ferrous Molding Projects No. 30, 31, and 32.
18	Non-ferrous Castings	222–251		
19	Casting Defects	Review	10	Non-ferrous Melting.
20	Design of Castings	253–263		
	Final Examinations			

[1] See references at end of paper.

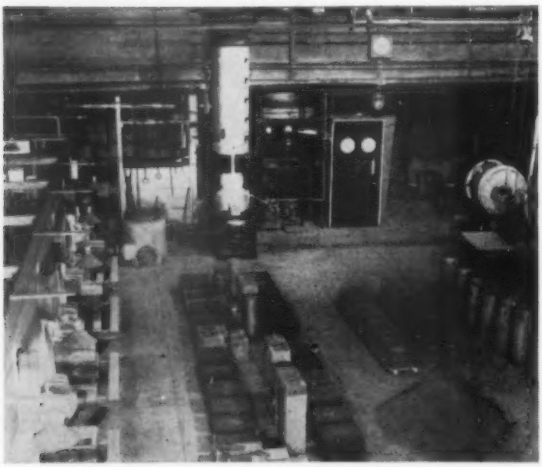


FIG. 3—STUDENT PROJECTS AND TEST BAR MOLDS READY FOR POURING.

of engineering makes it much easier to arouse and hold the students' interest. Problems of the local industry are discussed and solved in the classroom, and from this, the students realize that a scientific approach is important to a foundry problem. Students showing special aptitude in the elementary course are encouraged to take advanced work.

INTERMEDIATE FOUNDRY PRACTICE COURSE

10. Intermediate foundry practice is offered as an elective to sophomores who desire to continue their study of foundry practice. Each year, approximately 25 students take this course, which maintains their interest while they complete the chemistry and physics prerequisites for the more advanced foundry courses.

11. In this course, the students study the production of nonferrous metal castings, malleable iron castings, and steel castings, and the principles of sand testing and control. Inspection trips are made to local foundries to study processes which cannot be carried out in the university foundry laboratory.

FOUNDRY CONTROL METHODS COURSE

12. This course was introduced 3 years ago, after a university study of the future of manufacturing process courses revealed that the rapidly developing foundry industry needed engineers, control laboratory technicians, and executives. Since neither laboratory nor equipment were available for such a course, the teaching staff and the students built a laboratory and most of its equipment at a minimum of expense in the university shops.

13. A 30x30-ft. laboratory was built as a mezzanine floor over part of the foundry. Laboratory benches, exhaust hoods, and cabinets were built and installed. An 8-kva., multiple-tap, transformer and two silicon-carbide tube furnaces were made for secondary calibration of thermocouples; a smaller coil-wound resistance furnace was made for primary calibrations. Aluminum castings, poured and machined in the shops, were combined with a motor-driven agitator to make a rapid clay washer, and a discarded gyratory riddle was rebuilt for testing grain-fineness of sands. The foundry department also designed and built a furnace and accessories for making rapid sulphur determinations. An obsolete x-ray machine was secured and rebuilt to deliver 280,000 volts at 5 milliamperes. To protect operators from the rays, a special cubicle in one corner of the laboratory was lined with 3/16-inch lead sheet. Special equipment was installed so that radon gas, which is obtained from the Department of Chemistry, could be used in making gamma ray studies. A control room and developing room completed the x-ray division.

14. For convenience in teaching, and to conform with industrial practice, the foundry control methods course is divided into the following sections:

1. Chemical and mechanical control of foundry raw materials.
2. Chemical and mechanical control of processing materials.
3. Scientific control of manufacturing processes in the production of castings.
4. Inspection, calibration, and maintenance of foundry production control equipment.
5. Chemical, mechanical, and metallurgical inspection of the finished cast products.

15. Under the first section, the students learn some of the industrial methods of inspecting incoming raw materials such as pig iron, scrap iron, scrap steel, coke, limestone, molding sand, and core sand. Supervised by the teaching staff, the students analyze iron and steel for carbon, silicon, manganese, sulphur, and phosphorus content; coke for moisture, ash, volatile combustible matter, fixed carbon, and sulphur content; limestone for ignition loss, silica, "combined oxides," calcium oxide, magnesium oxide, and carbon dioxide; and sand for moisture, clay-content, and fineness. Fig. 4

shows some of the chemical equipment and students conducting tests on raw materials.

16. Control of processing materials involves the testing of molding sand for moisture content, effect of moisture, clay content, bond strength, fineness, and permeability. The testing and grading of the foundry sands are carried out according to the methods prescribed by the American Foundrymen's Association. Fig. 5 shows some of the sand testing equipment and students conducting tests on molding sands.

17. Operation by the students of a 15-in. cupola is one example of scientific control of a foundry maunfacturing process. This cupola was designed and built in the university foundry. It is equipped for maximum control during operation. Sized coke and metals of uniform composition are weighed accurately. Combustion is automatically controlled to maintain predetermined relative quantities of CO_2, CO, and N_2 at a certain height above the melting zone. This control is achieved by means of a device which automatically samples and analyzes the stack gases every 2 min. readjusting the blast gate when the quantities change. Both pressure and volume of the incoming air are measured and recorded. A gas-heated, silica-gel, dehumidifier dries the air for the blast.

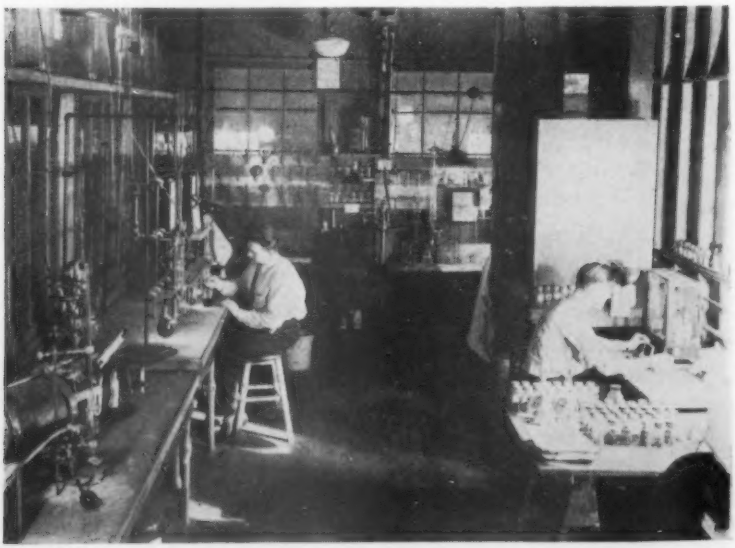


FIG. 4—STUDENTS ANALYZING RAW MATERIALS IN THE CONTROL LABORATORY.

FIG. 5—STUDENTS TESTING MOLDING SAND IN THE CONTROL LABORATORY.

18. The exact desired humidity is obtained by mixing dry and moist air in a suitable arrangement of ducts, and checked by means of wet and dry junction thermocouples located at the blower intake duct. A definite tapping sequence is maintained by stop-watch timing. The temperatures of the metal at the spout and at the mold are rapidly and accurately measured with a tungsten-graphite thermocouple[2*] perfected in the foundry department. Fig. 6 shows some of the control instruments used in connection with the 15-in. cupola and students conducting a test heat.

19. Another example of process control is illustrated in the methods used to make a test bar mold. Molds for test specimens poured in green sand are machine-made from synthetic sand No. 10[3]. This sand is prepared by rebonding the dry sand with bentonite and gelatinized starch, and mixing with water in a muller. The molds are jolted a definite number of times and squeezed at a specific pressure.

20. Similar control is studied in the non-ferrous melting and sand mixing departments, and in the core room.

21. Equipment such as pyrometers, pressure gauges, weighing scales, and flue analyzer, are calibrated periodically by the students under supervision. For example, thermocouples are calibrated by immersing in a small crucible of metal held in a silicon-

* Superior numbers refer to references at end of paper.

carbide tube furnace. The temperature of the molten metal is determined by a noble metal couple which has been checked against primary standards. Fig. 7 shows some of the equipment and students calibrating pyrometers.

22. The students study causes of casting defects in the control laboratory. X-ray and gamma-ray examinations are made of pressure castings.

23. Rapid routine methods of analysis are used to determine the carbon, silicon, manganese, sulphur, phosphorus, and alloy content of the test bars. A machinability test is conducted on test bars, as well as the common transverse, deflection, tensile, and hardness tests.

24. Test bars studied include transverse bar and tensile specimen, step bar, contraction bar, flowability specimen, and various chill specimens. The microstructure of the test bar is correlated with raw materials, processing, and resulting mechanical properties. Fig. 8 shows a student setting up a casting for x-ray inspection.

25. Study assignments for this course are made from the following:

Transactions of the American Foundrymen's Association; Sand Control, Scobie, 1940; *Testing and Grading of Foundry Sands,* A.F.A., 1938; *Alloy Cast Irons,* A.F.A., 1939; *Cast Metals Handbook,* A.F.A., 1940; *The Microscope in Elementary Cast Iron Metal-*

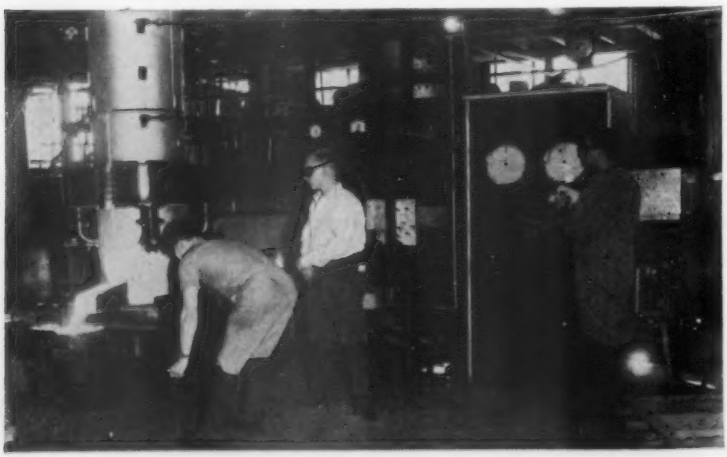


FIG. 6—STUDENTS RUNNING TEST HEAT ON A 15-INCH CUPOLA.

FIG. 7—STUDENTS CALIBRATING PYROMETERS IN THE CONTROL LABORATORY.

lurgy, Allen, A.F.A., 1939; *Industrial Radiography*, St. John and Isenburger, Wiley, 1934; *Chemical Analysis for the Foundry Control Laboratory*, Scobie, 1939; and certain publications of foundry suppliers.

FOUNDRY PROBLEMS COURSES

26. Two problem courses, Foundry Research Problems and Advanced Engineering Problems, are offered to upper-class men and graduate students. Students electing to take these courses assist in research problems sponsored by the foundry department.

27. The following are some of the research projects now being conducted:

1. Effect of moisture control of the cupola air supply on the properties of the iron,
2. Effect of combustion control in a cupola on the melting losses,
3. Torsion impact testing of cast iron,
4. Effect of furnace atmosphere on the porosity of pressure castings,
5. Synthetic molding sands,
6. Rapid temperature measurement of cast iron and steel. and
7. Rapid methods of chemical analysis for the foundry.

FOUNDRY CONTROL METHOD COURSE CONSIDERED A SUCCESS

28. The success of an elective course in a university curriculum may be based on the increase in enrollment in the course and in the number of students who find employment after graduation in the industry.

29. Figure 9 shows the increase in enrollment in the foundry control laboratory course and the two foundry problem courses. This rapid increase in enrollment has overcrowded the facilities which 3 years ago were considered more than adequate. Six of the students who took this course, and have obtained their University degrees, are now employed as foundry engineers, instructors of foundry practice, or research technicians.

MINNESOTA STUDENT CHAPTER OF THE A.F.A.

30. The interest the students have in the foundry practice courses is shown by the recent formation of the first student chap-

FIG. 8—STUDENT ADJUSTING CASTING FOR X-RAY EXAMINATION IN THE CONTROL LABORATORY.

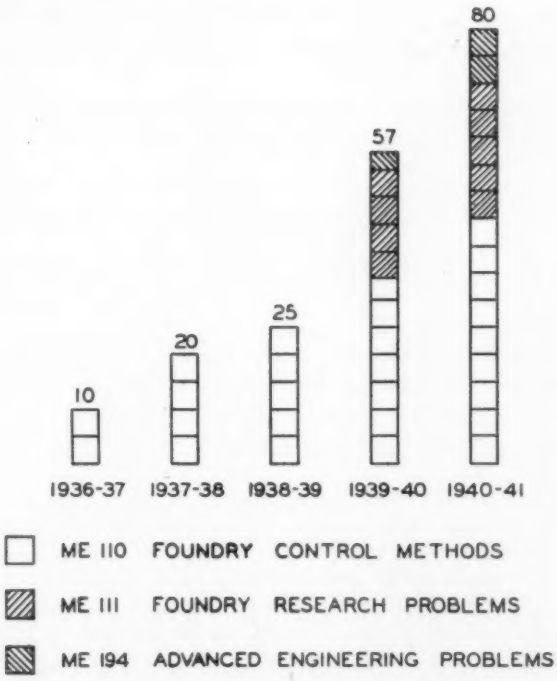


FIG. 9—REGISTRATION IN ELECTIVE ADVANCED FOUNDRY COURSES.

ter of the American Foundrymen's Association in the United States. The group is active in campus affairs and does much to promote interest in foundry education and the foundry industry.

ACKNOWLEDGMENT

31. The authors wish to thank Professors John R. DuPriest and Charles A. Koepke of the Mechanical Engineering Department and the many students who, by their assistance and cooperation, have made these courses possible.

References

1. Campbell, H. L., *"Metal Castings,"* John Wiley & Sons, New York, 1936.

2. Holtby, F., *"Rapid Temperature Measurements of Cast Iron with an Immersion Thermocouple,"* TRANSACTIONS A.F.A., vol. 47, p. 854 (1939).

3. Holtby, F. and Scobie, H. F., *"Recent Experiments with Gray Iron Synthetic Molding Sands,"* TRANSACTIONS A.F.A., vol. 48, p. 465 (1940).

DISCUSSION

Presiding: F. G. SEFING, International Nickel Co., New York, N. Y.

A. V. O'BRIEN[1] *(Written Discussion):* Foundry work has been justly characterized as an art, a science and an industry. As an industry it is the very basis of all manufacturing, in fact, of all engineering. However, foundry work in the engineering college curriculum has been generally viewed not as a revered parent but as a slightly defective stepchild. The authors of this paper are to be complimented on their work which cannot fail to improve the standing of this subject among engineering educators.

A reason for the low esteem in which the foundry has been held is that it was the *art* that was taught and not the *science*. The art has been practiced for centuries, but the underlying science has been imperfectly understood until comparatively recent years.

In none of the other shops which are traditionally a part of the engineering curriculum can the scientific moral be so aptly pointed. The chemistry of the combustion of fuels, the physics of gas and metal pressures in molds, the geologic origin and chemical composition of sands and refractories, the effect of design on the cooling rates and shrinkage of castings, as well as the chemical and physical properties of such auxiliary materials as core bonds and facings, all these, in addition to the most important of all, the science of metallurgy, have a definite bearing on the success of a foundry enterprise.

I agree entirely with the authors as to the futility of "melting scrap and pouring book ends," solely for the purpose of supplying souvenirs for the students. If the making of a book end provides some useful application of the art and science all well and good. We still have to "pound sand," make and set cores, vent, and use chaplets for these are all definitely a part of the art, and if our application of the art is not successful, no amount of science will save the situation. However, I do not favor the making of molds until a student can produce a perfect casting. There is really more educative value in a defective casting than in a good one, and after all, our boys are not going to be molders. Incidentally, the grading of students entirely on the perfection of their workmanship is decidedly unfair, in any shop course.

At the University of Iowa, the foundry course for sophomore mechanical engineers is a 2 credit course of 32 class periods of three hours each, of which one hour each period is devoted to lecture and recitation. Of the balance, about ¼ of the time is spent in the sand laboratory on the routine tests of molding and core sands, ⅛ on the core bench, ⅛ at furnace attendance, and the rest of the time making molds by hand and machine, for iron and non-ferrous castings, including the molding of test bars in both types of metal, which are broken in the materials testing laboratory, and the results presented in the form of a report.

[1] Superintendent, Manufactures Laboratory, University of Iowa, Iowa City, Ia.

Chemical and electrical sophomores have a somewhat similar course, necessarily condensed, since they have only one semester of shop work altogether.

A senior elective course for mechanical engineers for three credits, includes advanced sand testing, practice in the rapid methods of chemical analysis of the ferrous metals, preparation and micro-photography of specimens of the cast metals.

R. P. SCHNEIDER[2] *(Written Discussion):* We are very glad indeed to see such a paper given a place in the work of the American Foundrymen's Association. The foundry industry is rapidly becoming more and more scientific and it is up to the Universities to realize this new situation.

As the authors remark, the old days of learning foundry practices are about over. More and more, we are concentrating on the fundamental theories and practices behind the work of a foundry man in our work at Ohio State. Where we formerly gave one lecture and six hours of practical work per week we now give two lectures per week and cover much the same ground as listed in the authors' paper.

In our foundry we do not make castings for the fun of having them remelted. We make castings for use and throughout all the work the value of cost is emphasized, from the cost of the raw materials leading up to the cost of the finished product. This policy is not confined to the foundry, it is common to all of our shops. We endeavor to make our students cost conscious.

The foundry is not an individual, independent unit standing by itself, but is integrated into the work of other departments. The castings flow into the machine shop and there meet up with welded products and steel products, and the student can see and is told the relation of cast structure to other forms of raw material. This policy is pursued particularly with our welding engineering where their men have an idea that welding takes the place of foundry work. They quickly learn that they have serious competitors in the casting industry. In our course in Waste Elimination, Work Routing and Selection of Material the foundry offers us an excellent example of practical study. In other words, our foundry work ties in with our other courses.

We wish to congratulate the authors on their well integrated views of courses in foundry practices. We have only two. The first, concentrates on the practice and theory of casting and particularly of the metal side of the industry. The second course, rather more advanced, deals largely with the fundamental theories of sands used and the methods of testing them.

We believe with the authors that there is a distinct need for educational training in foundry work, a training which can be better given in the University than in Industry.

[2] Assistant Professor, Foundry Practice, Dept. of Industrial Eng., Ohio State University, Columbus, Ohio.

W. A. SNYDER[3] *(Written Discussion):* Messrs. Holtby and Scobie have developed foundry courses that in addition to teaching routine fundamentals are capable of turning out foundry technicians. It is important to note that advanced students can work on problems taken directly from industry or on research which may provide facts for immediate or future use by industry.

Control and research are the two most important factors in the foundry today. Application of these will bring down costs and give the engineer greater latitude in his design. Application of the courses discussed will certainly expand the possibilities of the foundry. There is a recognized problem, however, that cannot be solved in the foundry. This problem may be resolved into two parts; first the failure of engineers to recognize the full possibilities of cast metals, and second the failure of engineers to recognize the limitations of cast metals. The industry would benefit if misuse of cast metals was eliminated.

This problem may be met by the reorganization of foundry courses as suggested by the writers, from "sand pounding" courses to studies better suited to engineering students. This means more lecture hours and more time devoted to the design and physical properties of castings. Since most students come to elementary foundry courses with little or no knowledge of foundry work, the greater part of the lecture material even with increased periods must be devoted to fundamentals of foundry practice. It is therefore interesting to contemplate an intermediate course devoted to a more thorough study of design of castings and properties of cast metals. The laboratory work of this course would familiarize the student with the preparation of various cast metal test specimens. This would help him properly interpret test results when applying them to design.

An optional course of this nature following a well balanced course in the fundamentals of foundry practice might do a good deal towards preventing the misuse of cast metals.

E. J. ASH[4] *(Written Discussion):* The writers are to be congratulated on the preparation of this publication, first, because of the efforts being made to further the interest in the casting industry, and second, because of the increased interest shown by college students in electing advanced courses in the subject. One must conclude that the increasing enrollment in elective courses is largely a result of the enthusiasm created in the subject in the elementary or required course.

The subject matter to be taught in a course dealing with cast metals is always a debatable question. Specifically, how vigorously should the many and varied phases in producing cast metals be covered, and how much detail of practice and theory of fundamental principles should be given. How much, in a course of this nature, should be devoted to fac-

[3] Associate, Mechanical Engineering Shops, University of Washington, Seattle, Wash.

[4] Associate Professor, Dept. of Metal Processing, University of Michigan, Ann Arbor, Mich.

tors that may be classed as general education, and how strongly should the more practical and useful aspects be stressed.

That a thorough course in cast metals should be a part of an engineer's education is emphasized at the present moment. The beginning of our Defense Program was immediately felt in the foundry, indicating that this is one of the nation's basic industries. The foundries were called upon to produce the structures which in many instances were very essential items in the construction of production machines. The obvious conclusion to be drawn regarding the high estate of our foundry industries leads one to believe that insufficient attention is being devoted to this subject in our engineering schools. Not only should the course be required of metallurgical and mechanical engineering students, but also civil, marine, electrical, and, especially, aeronautical students as well.

The University of Michigan offers the following courses dealing with cast metals:

1. There is an elementary course required of all mechanical engineering students, consisting of thirty-two lectures and an equal number of laboratory periods. The combined lecture and laboratory are four hours per period, the lecture or recitation period absorbing approximately one and one-half hours.

2. A second course, very similar to the first, is required of metallurgical students and consists of only sixteen periods of four hours each as above.

3. A third course, elective, deals with foundry management and costs.

4. A fourth course, termed Advanced Foundry, is offered to students who wish to elect any particular study in which library and laboratory work are combined.

The following remarks are offered with regard to the nature, function and scope of a course in an elementary study of cast metals:

It is agreed that courses in foundry should not be sand-pounding courses. Laboratory work should consist of a judicious amount of molding, but only sufficient to permit the student to become acquainted with the difficulties involved in molding, and appreciate the necessity of proper molding materials and equipment. The writer feels that further exercises in molding is merely wasting the student's time. Educators should realize that students are not being fitted for molding as a life work. The students themselves are quick to recognize this fact.

One of the most important and useful phases of a course in cast metals is to give the engineering student a clear conception of the properties, uses, and limitations of the various types of cast materials available in our foundries. Amplifications of the above statement involves a host of factors. Emphasis should be placed upon the vast improvements possible in cast structures as a result of scientific control in melt-

ing and casting. It is believed that much of the advance in our cast products in the past two decades was due not so much to new and better methods of gating, but more so through application of basic metallurgical principles. The design of cast structures is very important to our engineering students. Very few students after leaving school enter our foundries as production men, but many are confronted in their fields of work with design of structures and selection of material. The poor design, with regards to the behavior of molten metal during solidification, which comes from drawing boards has in a number of instances seriously affected the prestige of cast metals in general. With a thorough understanding of the mechanism of solidification, designs can be translated into serviceable sound castings and do much to eliminate the impression that castings are not reliable because of internal discontinuities. The proper selection of cast material for a given application cannot be overemphasized. An apt slogan should be "Why not make the structure of cast metal."

Laboratory work should further acquaint the student with the various types of tests and testing equipment used to evaluate raw material and finished product. With regards to the technique of making tests, it should be borne in mind that this class of work can be over stressed. Any young man of average intelligence, without benefit of college education, can be made readily relatively proficient in making routine physical and chemical tests. Many employers wish to train their men in these lines of work in their own plants.

There are many other items to be stressed in a well-balanced course in the production and use of cast metal structures. Laboratory work is necessary, but a thorough understanding of fundamental principles underlying the production of serviceable castings which can only be given in classroom discussions is very essential. Instructors should bear in mind they are training men who will, in general, specify and use castings, rather than make them.

C. H. CASBERG[5] *(Written Discussion):* The authors are to be congratulated on the preparation of the paper, on the work required to set up a new foundry course, and on securing the consent of the administrators of the university to offer a new course.

Some may think that it is very easy to set up and to offer a new course in a university. It is not a simple matter because the administrators of a university scrutinize a proposed new course just as carefully as does the executive in any business scrutinize the production of a proposed new product. I have not seen or talked to these men, but think in getting permission to offer this new course, they had to show that there was need for this course, that a sufficient number of students were interested enough to register for it, and that the expense per student would not be greater than other courses of this type—as well as to answer numerous other questions. Their enthusiasm probably caused

[5] University of Illinois, Urbana, Ill.

them to promise to teach the course in addition to their present load without additional compensation. In other words, their enthusiasm probably got them a lot of hard work, and no doubt, they wondered sometimes whether or not it was all worth while.

I believe that there is much need for courses of this kind in other universities, but if courses of a similar nature are set up, it is necessary to have instructors with enthusiasm and willingness to work, to have support from the outside, such as the A.F.A. and the executives in the foundry industry.

At the University of Illinois at the present time, we do not offer advanced courses in foundry engineering, although we are much better prepared now than we were ten years ago. We have had some help and encouragement from the foundry industry, from manufacturers of foundry supplies and equipment, and from the A.F.A. This help and encouragement from sources outside the university have been enough to encourage the instructors in foundry work so that they think their job is important and put some enthusiasm into their work.

DONALD S. CLARK[6] *(Written Discussion):* Apparently the object of the courses set up at Minnesota is to familiarize engineering students with problems confronting the designer, and the limitations as well as the procedures in the foundry. It is interesting to note that this work is sponsored by the Mechanical Engineering Department. This is further evidence that the engineering colleges believe that the designer must have a clearer understanding of the possibilities and limitations of production procedures.

The authors do not state what preliminary training the students have had prior to taking the Elementary Foundry Practice Course other than Chemistry and Physics. The text for this course is "Metal Castings" by Campbell. This is a very good book for this course provided the students have received proper instruction in some of the more basic phases of metallurgy which are not included in this text.

Because the curriculum of the engineering school is already crowded, it seems most desirable to give all possible emphasis to the basic studies before there is much specialization. The engineering schools are tending to decrease the amount of shop work because they have come to realize that they are not training men to become shop mechanics. All engineers should have some familiarity with shop processes because without it they cannot make sound economical designs. However, it seems that the course would be more effective if it were given somewhat later in the curriculum.

It should be the objective of any course of this type to use the laboratory, shop or foundry as a place where the basic principles which have been discussed in class can be demonstrated and applied. As long

[6] Assistant Professor of Mechanical Engineering, California Institute of Technology, Pasadena, Calif.

as the course serves this purpose and does not develop into a technique training course, it should prove to be of inestimable value.

PETER E. KYLE[7] *(Written Discussion):* The authors have presented a story of their experiences in organizing a course in foundry control methods which undoubtedly can be classed as one of the best being offered in this country at the present time.

Such a paper should give the schools a very good idea of how much can be done with limited funds. It indicates that if emphasis is placed on fundamentals it is not necessary to have the most elaborate shop geared for production with all models of the most modern foundry equipment.

The extent to which chemical analysis, metallography, and mechanical testing can be taught in the foundry courses will of course vary among schools. A close coordination with foundry work should be had if these subjects are treated by other departments.

Arousing student interest in the subject seems to be one of the most important and difficult problems to solve. The authors should be congratulated on the progress they have made in this respect.

We have every indication that good courses in the field of foundry practice and engineering are being developed in the engineering colleges. In most cases educators are in accord with the accepted standards such a course should meet but quite often their funds and staff are inadequate. As soon as these conditions improve we need not fear for the quality of the work which these schools will undertake. Reviewing the work being done at a number of schools should be of help in keeping the educational institutions informed of the progress being made.

DR. J. F. OESTERLE[8] *(Written Discussion):* The authors of this paper are to be congratulated on a very interesting presentation of new methods of bringing the foundry to the attention of the engineering student as a professional opportunity.

To us who deal more definitely with metallurgical engineering, these new courses in foundry practice attain a new aspect. With previous interest in foundry shown primarily as an element in the general scheme of production methods, the broadening effect of the new approach to this professional problem as shown in the paper is delightfully refreshing. With more emphasis on control and the underlying scientific principles, it seems that one would find a demand for advanced courses.

Courses in foundry practice attract the engineering student from many viewpoints. Every engineer likes to do things with his hands—the old college foundry satisfied this desire. But after the doing, there must come the final satisfaction of understanding the principles involved.

---

[7] Asst. Prof. Mech. Engineering, Massachusetts Institute of Technology, Cambridge, Mass.

[8] Professor of Mechanical Engineering, University of Wisconsin, Madison, Wis.

In addition, melting metals for the first time and pouring them has a fascination second to none. What the authors have shown to be the need at Minnesota will without doubt be found to be the need at every other engineering college.

University courses in foundry control methods should interest every engineer. Castings serve without competition in many machines, and castings are certainly holding their own in national defense. Castings offer the opportunity to study machine design, physical properties, metallographic structural properties, chemical composition and many other of the important physical and chemical features of metals.

All foundry courses at Wisconsin are now administered by the department of Mining and Metallurgical Engineering. It was turned over to this department in order that the theoretical metallurgical aspect might be more properly developed. Each administering department naturally emphasizes the viewpoint it thinks to be most important. As metallurgical engineers it is our opportunity to point toward the production of sound, non-segregated, fine-grained castings. And we cannot lose sight of the fact that design, while important to the user of castings, is also our concern. The variables in foundry practice that constitute the foundry problems are numerous. It is only by the intelligent understanding of those factors that contribute to a perfect casting that the engineer can administer the foundry; and it is also true that a foundry that is intelligently administered is a foundry with a high percentage production of good castings.

It might be well to conclude with the thought that what is one man's headache is another man's pleasure. When all the variables in a foundry are fairly well understood it is the job of the engineer to keep them synchronized. It is also the aim of the American Foundrymen's Association through many annual foundry conferences to help synchronize these variables.

MESSRS. HOLTBY AND SCOBIE *(Authors' Closure):* The authors wish to thank those from the other universities for their interesting discussions of this paper. These written discussions include many of the questions and problems which had to be answered and solved before our courses were developed and accepted. The belief that advanced courses in foundry practice has an important place in an engineering curriculum is substantiated by this response and by the increased number of such courses taught.

The courses discussed in this paper were used as a foundation for a national defense training course in "Foundry Control and Casting Inspection." Thirty men successfully completed this course and are now employed in the foundry industry as inspectors, foundry engineers, or foundrymen. These men, although many of them were working full time in industry, attended the University 3 hours each night, five nights each week for 24 weeks, or 360 hours. In addition they were required to study 120 hours on outside assignments.

The objective of this course was to train a number of select 2-year college men to be foundry control technicians and casting inspectors. Lectures given by a number of University faculty members covered such fields as general foundry practice, sand testing and control, melting operations and control, physical material testing, chemical analysis of raw material and finished product, pyrometry control, x-ray inspection of castings, macro and micro-inspection, casting defects and causes.

A close correlation was maintained between the theory discussed in lectures and the actual laboratory work. Typical problems, such as found in the foundry industry, were studied and solved. The course material covered in the 24 week period was equivalent to one year of college education.

New material and equipment which was developed for this defense course is being incorporated in the courses described in this paper. The authors feel that this material as well as the experience gained in teaching men from the foundry industry will result in further improvement of these courses.

Some Causes for Porosity and Leakage in Non-Ferrous Castings†

By Alfred H. Hesse*, Washington, D. C.

Abstract

The causes for porosity and leakage in non-ferrous castings are discussed. Gas porosity and intercrystalline shrinkage porosity are studied in detail, including the effect of various gases, the influence of alloying elements, and the influence of pouring temperatures on porosity. The influence of pouring temperature on the porosity of a modified tin bronze alloy is also studied. An attempt is made to establish a relationship between per cent soundness and leaky castings.

Introduction

1. Non-ferrous copper-base alloys find many applications in industry. Among these alloys, one finds the tin bronzes used rather widely for many purposes, and by delving still farther, one finds pressure castings as one of their important applications. In many foundries, particularly jobbing foundries, the percentage of rejections due to leakage or porosity is sometimes quite high. These rejections are obviously very costly, since leakage is not generally found before machining. It is with a view to determining some of the causes for leakage in pressure castings and to devise methods for its prevention, that the work described in this paper was undertaken. The subject is not covered completely, since some investigations are in progress, the results of which will be reported later. In this work, no effort has been made to meet any specification requirements and the research was conducted to determine the effect of various gases, pouring temperatures, and compositions on porosity of castings and their physical properties.

* Division of Physical Metallurgy, Naval Research Laboratory.

† Published by permission of the Navy Department.

Note: This paper is a thesis presented by the author to the faculty of the School of Mines and Metallurgy, University of Missouri, Rolla, Mo., in partial fulfillment of the requirements for the degree of Metallurgical Engineer.

Note: This paper was presented before a Non-Ferrous Session at the 45th Annual A.F.A. Convention, New York City, N. Y., May 13, 1941.

2. There are several non-ferrous alloys commonly used for the manufacture of pressure castings. Two of these alloys seem to be more extensively used than the others, and for the purpose of simplification, they shall hereafter be referred to as tin bronze and leaded tin bronze. Actually these alloys have many trade names; for example, tin bronze is sometimes called composition "G" bronze, gun metal, 88-10-2, 88-8-4, etc., and leaded tin bronze, composition "M," steam or valve bronze, commercial 88-10-2, etc. The nominal chemical composition of the tin bronze referred to in this paper is 87.25 per cent copper, 9 per cent tin, 3 per cent zinc, and 0.75 per cent nickel. The nominal chemical composition of the leaded tin bronze is 88.0 per cent copper, 6.5 per cent tin, 4.0 per cent zinc, 1.5 per cent lead, and 1.0 per cent nickel permissible. These alloys are somewhat similar in chemical composition. Although the lead contained in the leaded tin bronze increases machinability, the tensile strength is lowered by the lead in combination with a lower tin content. However, from the standpoint of cost of materials, the leaded tin bronze is less expensive.

REVIEW OF THE LITERATURE

3. The lack of pressure tightness may arise from a number of causes, such as cracks, cold shuts, blowholes, and sand inclusions All these faults can be avoided by proper foundry technique. However, gas pockets and minute intercrystalline cavities or channels are more difficult to eliminate.

Gas Porosity

4. The possible sources of gas porosity are: (1) Absorbed gases may be present in the raw materials or absorption may take place during the melting process; (2) insoluble mold gases may be trapped in the metal during casting; and (3) metallic oxides present in the molten metal when cast, may be reduced to form carbon monoxide and carbon dioxide which are trapped during solidification.

5. The solubility of some gases in molten bronze increases with temperature, and during solidification the solubility decreases. Thus it is possible for any precipitated gas to become trapped internally to form gas cavities.

6. Investigations have been made on the effect on porosity of various gases which might be carried in with the raw materials

or absorbed during the melting process. Daniels[6‡] stated that carbon dioxide, nitrogen, and carbon monoxide have no effect on tin bronze and that hydrogen causes unsoundness. Jenkins[18], in his review of the literature on tin bronze, suggested that sulphur dioxide, and carbon monoxide are the occluded furnace and melting gases which form cavities. Bolton and Weigand[5] ran a practical foundry test involving over 100,000 castings (tin bronze and similar alloys) and found that, by melting under a neutral or slightly oxidizing atmosphere, a greater number of pressure-tight castings were produced than under a highly reducing atmosphere averaging 7 per cent carbon monoxide. Daniels[6] and Jenkins[18] confirmed the work of Bolton and Weigand[5].

7. Bailey[19] suggested that hydrogen and water vapor also might have a deleterious effect on pressure castings. Claus and Bauer[17] studied the influence of gas solubility on inverse segregation in tin bronzes and concluded that neutral gases, carbon monoxide plus carbon dioxide and nitrogen have a normal influence, while hydrogen gives extreme inverse segregation; that hydrogen sulphide, which reacts to form cuprous sulphide and hydrogen, has the same effect as hydrogen; and that sulphur dioxide, which forms cuprous sulphide and stannic oxide, has no effect. They found that phosphorus and lithium tend to reduce the dissolved uncombined hydrogen and decrease the segregation while zinc and aluminum have little influence. High internal gas pressure, developed by the evolution and entrapment of dissolved gases during solidification, was considered to be the cause of inverse segregation.

8. Bailey[19] showed data proving that the hydrogen solubility in molten copper depends not only on the temperature, but also upon the oxygen content of the copper. In other words, the higher the oxygen content, the lower the hydrogen solubility. There is a maximum oxygen content, however, which, if exceeded, lowers the hydrogen solubility but little. When the molten charge contained hydrogen, a greater degree of unsoundness was observed in sand castings than in chill castings; the apparent explanation is that a chill casting retains more of the gas in solution than a sand casting.

9. Only those absorbed gases which are soluble in molten bronze and whose solubility increases with temperature are to be considered as sources of trouble. Gases absorbed during the melting

‡ Superior numbers refer to bibliography at end of paper.

process, which are soluble in bronze at room temperature, naturally will not cause unsoundness or porosity. And, it seems highly improbable that insoluble gases could be held in suspension in molten bronze because of their buoyancy in liquids.

10. *Sources of Mold Gases.* Mold gases, which are trapped during the casting and solidification process, may come from a number of possible sources, for example, steam from green sand, air held in the pores of the sand, gases from the interstices of chills, carbonic oxides originating from ignition of carbonaceous mold facing materials and core compounds, and gaseous hydrocarbons from the decomposition of organic compounds[18]. Some of these gases are dissolved, but most of them remain insoluble in the form of distinct bubbles. If the metal is too viscous to allow the gases to escape, they will be retained in the form of blow holes distributed irregularly throughout the casting, but often confined to the surface[7].

11. The formation of tin oxide is considered by some writers to be harmful in tin bronze and particularly in leaded tin bronze. The presence of tin oxide usually is accompanied by porosity and low tensile strength. It seems probable that the metallic oxides are reduced by dry core binding material and produce oxides of carbon which are trapped with the freezing metal around the core, under the surface of the casting[18]. This trouble is encountered more frequently in bronzes of high lead content[11].

Intercrystalline Shrinkage Porosity

12. Some writers feel that intercrystalline shrinkage porosity is caused by slow or prolonged freezing with consequent sluggish feeding of sections. The copper-tin equilibrium diagram shows an extremely wide freezing range of approximately 150°C. (302°F.) for a tin content of 9 per cent. During solidification, bronze contracts considerably. Thus, the dendrites which are formed, occupy less space than the liquid from which they are formed and voids are left around them. If new liquid is available during solidification, the spaces tend to fill, but if, for any reason, the supply of liquid is cut off, contraction cavities will result. It is obvious that, due to the wide cooling range, a considerable depth of metal from the cooling face of the casting will be freezing at the same time and the feeding liquid will have to traverse a long network of dendrites

to fill the cavities[7, 16]. Bolton and Weigand[5] stated that the gas rejected in the crystal interstices during crystallization prevents feed back of the metal and causes shrinkage by preventing feed through back pressure. Dews[7] stated that cavities are formed primarily by contraction and, while they may provide lodging places for liberated gases, they are not formed through any action of the gases.

13. It seems logical that, even in the absence of dissolved gases, intercrystalline shrinkage cavities may be formed by contraction during solidification, and that intercrystalline shrinkage porosity may be controlled by the rate of solidification. This seems to point to the importance of casting temperature; for example, if the casting temperature is well above normal, the contraction during solidification will be greater than if cast at a normal temperature. Consequently, if it is impossible to feed the contraction cavities with liquid metal, the unsoundness of the resultant casting will be increased. Karr and Rawdon[2] and Carpenter and Elam[3] showed that casting temperature has, through its influence on rate of solidification, a greater effect on soundness of tin bronze than any other factor. Primrose[1] and Thews[15] concluded that soundness, ultimate strength, and degree of elongation depend upon the rate of solidification. Smalley[4] showed that casting at too low a temperature results in poor physical properties which may be attributed to entrapped oxides, gas, and improper feeding.

14. *Minor Factors.* Since this paper is primarily concerned with the causes for porosity in tin bronze and with methods of attack for the elimination of porosity, no attempt is being made to discuss in detail those factors which have only a minor influence on porosity. However, it seems worthwhile to mention some of the methods employed to produce satisfactory castings. The castings are designed to maintain a cast finish where possible and to allow for a minimum amount of machining when machining is necessary. This procedure does not reduce porosity, but it does take advantage of the pressure-tight skin on the surface of the casting chilled by the mold. From the standpoint of design and molding, there are several important factors to be considered; for example, the coring of heavy bosses, the maintenance of uniform thickness throughout, the use of ample fillets when changes in sections or right angle bends are necessary, the selection of the proper type and size of gates, and the proper use of chills[9, 18].

Careful selection of raw materials and scrap are made to avoid possible contamination by undesirable elements, gases, or oxides. Melting and casting procedures are carefully controlled, and permissible elements, such as nickel[10,13] and phosphorus[7,8,14] are added. These alloying elements are considered beneficial to the production of pressure castings when properly used.

EXPERIMENTAL PROCEDURE

Study of Gas Porosity

15. The literature, in general, agrees that reducing gases are a source of porosity in tin bronze castings, although there is a difference of opinion in regard to the gas or group of gases which may be contained in the raw material or scrap, or which may be absorbed during the melting process and evolved during solidification. In view of this, the following experiments were performed to determine the effect on porosity in tin bronze of the various gases which might be present in the atmosphere during melting.

16. *Zinc Losses Compensated.* Some experimental melts were made in vacuo to determine the feasibility of using virgin metals in the proper proportions for tin bronze. Considerable difficulty was experienced in retaining the zinc in the molten metal. The zinc, which vaporized, was carried off through the vacuum system and was deposited along the way as it became cool enough to condense. To avoid this loss of zinc, the following procedure was devised, which proved to be quite satisfactory for the entire series of experiments.

17. A preliminary melt, which contained more than the desired amount of zinc, was sand cast. Three-pound specimens, cut from the preliminary melt, were vacuum melted in a graphite crucible and remelted in a graphite crucible under certain controlled gas atmospheres. A different specimen was melted under each of the following gas atmospheres: hydrogen, water vapor, air, sulphur dioxide, carbon dioxide, nitrogen, oxygen, and carbon monoxide. The excess zinc was purposely added to the preliminary melt to compensate for zinc losses which occurred during vacuum melting. It is to be noted that the specimens cut from the preliminary melt were melted in vacuo before melting under controlled gas atmospheres. This procedure was followed because it was felt that the sand cast specimens might contain certain gases or oxides which would influence results when melting under controlled gas

atmospheres. In other words, by following this procedure all specimens were brought to a standard quality and thus the results obtained after melting under different gas atmospheres were comparable.

18. *Specimens Used.* The preliminary melt was made of virgin metals in a high-frequency, lift-coil, induction furnace using a No. 60 semi-conducting crucible having a capacity for 180 lb. of bronze. The metal was cast at 1140°C. (2084°F.) into a green sand mold. The resultant casting weighed approximately 117 lb., including gates and risers. The main body of the casting consisted of three cylinders, each 1¾-in. in diameter and 30-in. long. The casting was sand-blasted, thoroughly cleaned, and cut into lengths each weighing about 3 lb. The chemical analysis of a representative sample of the alloy showed that it contained 85.52 per cent copper, 8.71 per cent tin, 0.75 per cent nickel, and 4.73 per cent zinc.

19. To arrive at the desired zinc content (3 per cent), the specimens were weighed before and after vacuum melting and the loss in weight taken as loss of zinc. By this method, the approximate zinc content could be calculated. The charge was remelted if the zinc content was too high and discarded if too low.

20. *Melting in Vacuo.* Melting in vacuo, and under controlled gas atmospheres was done in graphite crucibles machined from Acheson graphite, electric-arc furnace electrodes, 4-in. in diameter. The crucibles were 2½-in. outside diameter, 5-in. overall length, 2-in. inside diameter at the top, tapered 4½-in. in length to 1¾-in. inside diameter at the bottom. A graphite hopper of 5-in. overall length was placed over the crucible so that the entire charge could be loaded at once, thereby preventing interruptions when melting. A graphite lid was placed over the hopper. Both the lid and the hopper also served to eliminate drastic temperature gradients during solidification. The crucible, hopper, and lid were placed in a quartz tube closed at one end, 12-in. long, 3½-in. outside diameter, and ⅛-in. wall thickness. A water-cooled bronze cap was fitted to the open end of the quartz tube and sealed with a rubber gasket. The cap had two convenient openings, one of which was closed when melting under vacuum. When melting under gas atmospheres, one opening was used as an inlet and the other as an outlet. A high-frequency, induction furnace was used for melting and a two-stage, rotary, oil pump was used to produce the vacuum. The total time for a vacuum melt was 60 min.—30

min. for evacuating the furnace, 25 min. to melt the charge, and the charge was held 5 min. in the molten condition after which the power was cut off and the metal allowed to cool to room temperature in vacuo.

21. *Melting Under Various Atmospheres.* The procedure for melting under gas atmospheres was as follows: Commercial grades of hydrogen, sulphur dioxide, carbon dioxide, nitrogen and oxygen were employed. To insure their purity, the gases were chemically treated to remove those impurities which might influence results. Hydrogen was passed over palladiumized pumice at 100°C. (212°F.) to convert oxygen to water vapor. It was then passed through concentrated sulphuric acid and calcium chloride to remove the water vapor before entering the enclosed heating chamber. Sulphur dioxide received no special treatment. Carbon dioxide was passed over calcium chloride. Nitrogen was passed over copper filings at 600°C. (1112°F.), through concentrated sulphuric acid, and over calcium chloride in the order named. Oxygen was bubbled through a cuprous chloride solution to remove carbon monoxide, and then over calcium chloride to remove water vapor. Water vapor was produced by running steam generated by boiling water through the heating chamber. Carbon monoxide was produced by continuously dropping concentrated formic acid into concentrated sulphuric acid. Any acid fumes and water vapor carried with the carbon monoxide were removed by passing through concentrated sodium hydroxide and calcium chloride. In all cases, excepting the melt in air, the exhaust gas was bubbled through water to maintain a constant pressure and to prevent suction of air into the system should a drop in internal pressure occur.

22. The flow of gas was maintained for at least 30 min. before starting the furnace and was continued until the metal cooled to room temperature. The metal was held molten for 30 min., reaching a maximum temperature of approximately 1100°C. (2012°F.). It solidified in about 15 min. and cooled to room temperature in 3 hours. The resultant specimens were halved lengthwise through the diameter, polished, and photographed. Density determinations and chemical analyses were made.

Study of Intercrystalline Shrinkage Porosity

23. *Effect of Alloying Elements.* In studying gas porosity, cavities were observed even in specimens melted in vacuo prior to

melting under controlled gas atmospheres. Assuming that by melting in vacuo no gases would be present, it seems reasonable to deduce that some other factor caused the porosity. Therefore, it was felt that this type of porosity might be natural for cast tin bronze and that some alloying element or elements were the contributing factors.

24. To prove this, the following experiments were performed. Five vacuum melts were made using the set-up described in paragraph 20. Virgin metals of high purity were used for a 3-lb. charge of each composition shown in Fig. 8. It is to be noted that in each successive charge copper is substituted for by one or more elements comprising tin bronze and that the last charge is the alloy itself. The specimens were processed in the manner outlined in paragraph 20.

25. *Influence of Pouring Temperatures.* The importance of pouring temperatures, often referred to as casting temperatures, was emphasized in the literature. Some experiments were undertaken to confirm statements in the literature, while others were made to determine the relationship between pouring temperatures and soundness, tensile strength and elongation.

26. Five heats, numbered from 22 to 26, inclusive, were run. Two were tin bronze containing no nickel, two were tin bronze containing the nominal nickel percentage, and one was a modification of tin bronze in which 6 per cent nickel was substituted for 6 per cent tin. The composition of each heat is shown in Fig. 11. The melting was done in a lift-coil type induction furnace having a capacity for a No. 70 crucible. Seventy-pound charges were melted in a No. 40 crucible. Each charge consisted of 30 per cent scrap and the balance virgin metal. Nickel was added in the form of a 50 per cent copper and 50 per cent nickel master alloy. The scrap metal, copper-nickel master alloy, and virgin copper were melted first. The molten metal was covered with charcoal and heated above the initial pouring temperature. The tin was added, followed by zinc. The crucible was removed from the furnace, the metal cooled to the desired pouring temperatures, and cast.

27. Details of the casting are shown in Figs. 1 and 2, Top. It was felt that results of tests achieved from this particular type of test casting would more nearly represent the properties of commercial castings than would results obtained from some of the

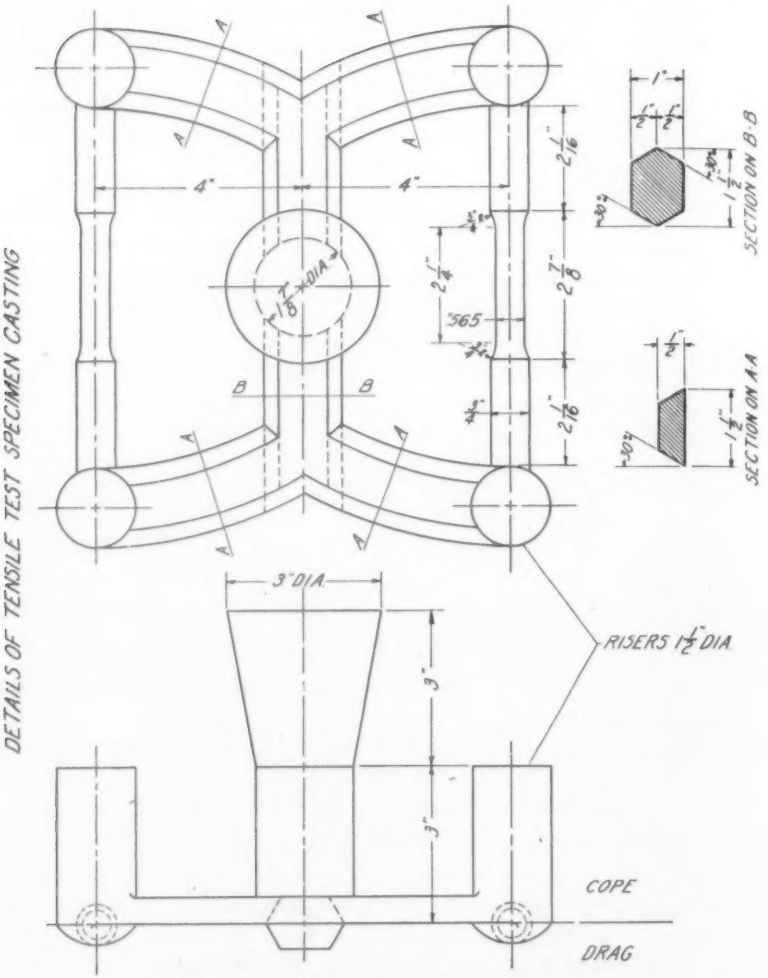


FIG. 1—DETAILS OF TENSILE TEST SPECIMEN CASTING.

other heavily gated types of test castings. The molds were made of No. 0 Albany green sand containing from 7 to 8 per cent moisture. The tensile test specimens were cut from each casting and density determinations were made on each specimen before machining to 0.505-in. diameter.

Hydraulic Pressure Testing

28. An attempt was made to correlate pressure tightness with

density by first determining densities of hollow cylinder castings as cast and as machined to size and then subjecting the machined castings to hydraulic pressure. Two melts were made in a No. 60, semi-conducting crucible, using the same equipment and following the same melting and casting procedure outlined in paragraph 26. Tin bronze containing no nickel was used. The details of the hollow cylinder castings are shown in Figs. 2 and 3. The

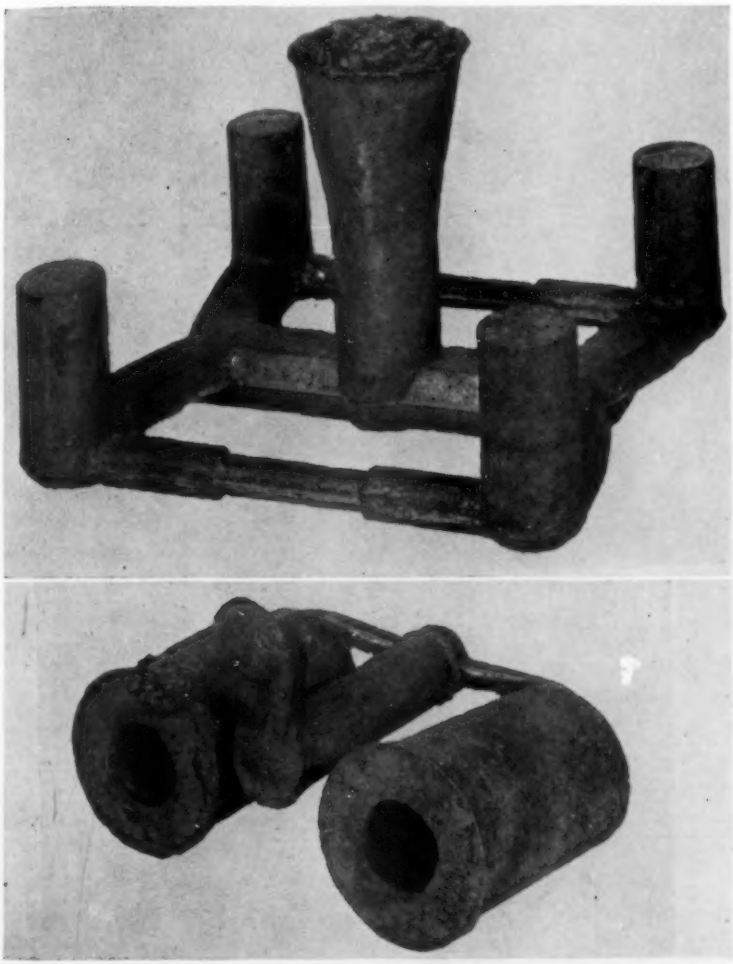


FIG. 2—(TOP)—TYPICAL TENSILE TEST SPECIMEN CASTING.
(BOTTOM)—HOLLOW CYLINDER TEST CASTING.

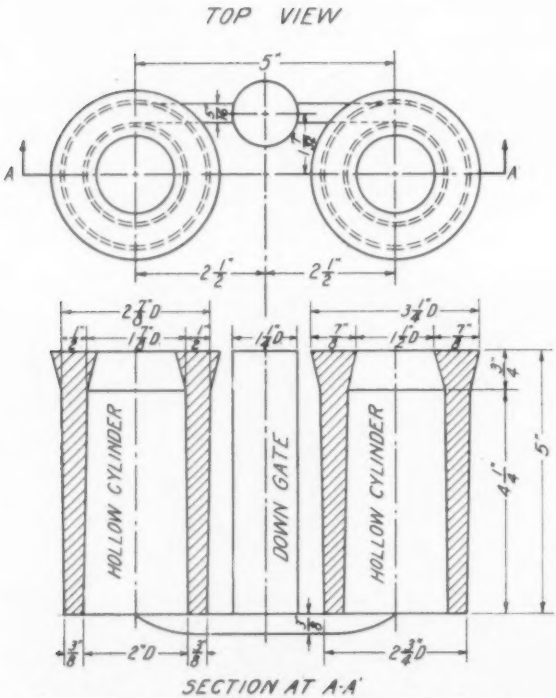


FIG. 3—DETAILS OF HOLLOW TEST CYLINDER TEST CASTING.

hollow cylinders were machined to 2½-in. outside diameter, 2¼-in. inside diameter, and 4-in. length for testing, the excess length being cut off the top section of the cast cylinder. Figure 4 shows the device used for hydraulic pressure testing.

Density Determinations, Voids and Soundness Calculations

29. Observed density values were determined by the conventional displacement of water method. The values obtained were at 20°C. (68°F.). It was found that densities of small samples taken from various sections of a casting varied considerably. Therefore, in all cases observed densities of the entire specimens were measured. In other words, in determining the density of a tensile test specimen the whole specimen was weighed in air and in water and the density computed. This gave an average observed density value for the tensile test specimen. The same procedure was followed in density determinations of hollow cylinder castings, and specimens obtained from vacuum and controlled gas atmosphere experiments.

30. To calculate per cent voids or per cent soundness, it was necessary to determine the maximum densities of the alloys studied. Specimens of each alloy were cut into one-inch cubes, and cold compressed and annealed until maximum and constant density values were obtained. Three methods of determining density were used: (1) displacement of water, (2) weighing a known volume in air, the volume determined by measuring the dimensions with a micrometer caliper, and (3) x-ray diffraction. The results obtained from the latter were used, since they were more consistent than the results obtained from the others. Table 1 lists the alloys studied and the corresponding maximum density values.

31. Before leaving the subject of density determinations, a few words should be included in regard to density as a measure of porosity or unsoundness. If a casting or specimen has a density equal to the maximum density, it can be said that the specimen is 100 per cent sound or contains zero per cent voids. On the other hand, if the density is less it is obvious that the specimen is not 100 per cent sound and necessarily contains a certain percentage

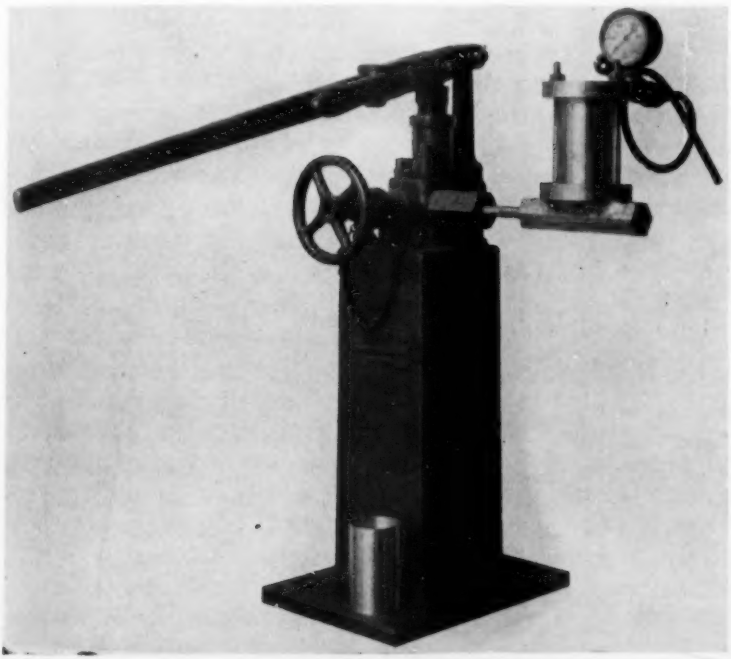


FIG. 4—HYDRAULIC PRESSURE TESTING DEVICE.

Table 1

MAXIMUM DENSITY VALUES

Chemical Compositions				*Max. Density X-Ray Diffraction Method, grams/c.c.*
Cu., Per Cent	*Zn., Per Cent*	*Sn., Per Cent*	*Ni., Per Cent*	
87.4	2.9	9.6	0.0	8.928
87.4	3.1	9.5	0.0	8.928
85.8	3.3	10.2	0.7	8.900
87.4	2.9	9.1	0.6	8.922
88.2	2.8	3.1	5.9	8.928

of voids to make up for the volume occupied. In other words, the observed density of a specimen is an indication of its soundness or porosity. The maximum density varies, of course, with the composition. Thus in presenting data showing the effects of variations in composition or the effects of other variables on porosity, observed density values must be compared with the maximum density obtainable for the particular composition. The percentage of voids may be computed from the following formula.

$$\text{Voids, per cent} = \frac{D_m - D_0}{D_m} \times 100$$

Where D_m is the maximum density and D_0 is the observed density as described in paragraph 29. The figures shown as soundness (per cent) were determined by subtracting voids (per cent) from 100. Thus a perfectly sound casting would be 100 per cent sound, and a casting containing 10 per cent voids would be 90 per cent sound.

DISCUSSION OF RESULTS

Study of Gas Porosity

32. Fig. 5 represents the data obtained from the study of gas porosity described in preceding paragraphs. Hydrogen and water vapor caused the most porosity, sulphur dioxide and air followed with a decided drop in porosity, and carbon dioxide, nitrogen, oxygen, and carbon monoxide caused porosity in the order given, with carbon monoxide causing the least. It is interesting to note the relative positions of the two bars marked "Leaky Castings." They represent hollow clyinder castings having the least porosity of 24 tested which leaked when subjected to the hydraulic pressure test. In other words, a further reduction of porosity seems necessary to obtain pressure-tight castings.

33. This discussion covers only the possibility of porosity caused by gases absorbed from the materials or during the melting process and precipitated during solidification. There is another type of gas porosity which may occur during the casting and solidification processes. This is in the form of occluded gases which appear near the skin of the casting only and are probably a result of insoluble gases formed by hot metal contacting the sand mold surfaces and which are trapped during solidification. These may be almost any of the gases discussed, and although such blowholes are undesirable, they are not too serious a defect if other factors

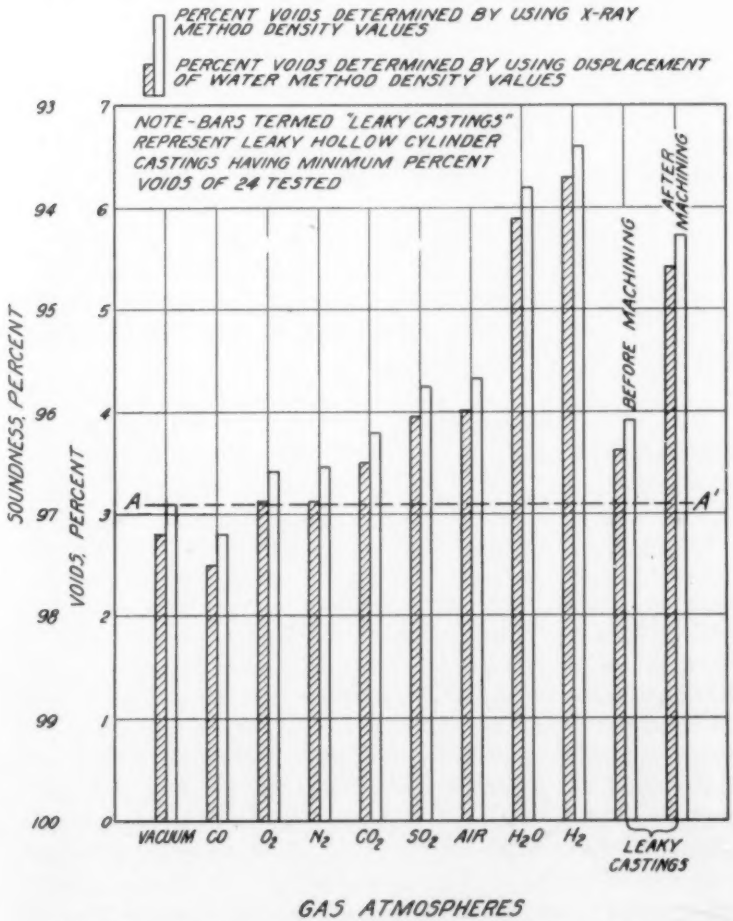


FIG. 5—EFFECT OF MELTING AND SOLIDIFYING TIN BRONZE UNDER VARIOUS GAS ATMOSPHERES ON SOUNDNESS.

governing porosity are properly controlled, since this type of porosity is usually found near the casting surface only and not throughout the casting. Thus, if the remainder of the casting is sound, it would be difficult to force liquid through the casting if the channels formed by a series of cavities are absent.

34. Porosity found throughout a casting might result from precipitated gas which was dissolved prior to casting, and obviously can cause leakage, particularly if the porosity is of such magnitude to provide convenient channels for the liquid or gas. It is interesting to consider in detail some aspects of absorbed gas porosity which are included in Fig. 5. The melt represented by the bar marked "Vacuum" (melted only under vacuum) can be assumed to be free of gas. By melting under vacuum in a graphite crucible, metallic oxides would be reduced to oxides of carbon, and these gases and others, if present, would be removed through the vacuum system. Even though the "Vacuum" melt is gas-free it has a certain amount of porosity (indicated by the per cent voids or per cent soundness of the metal) due to some other cause. Since the specimen was prepared under carefully controlled conditions all variables, including gas, were eliminated. Therefore, the other cause for porosity can only be attributed to intercrystalline shrinkage which will be discussed later.

35. If it is assumed that "Vacuum" has zero gas porosity, then a line drawn parallel to the horizontal axis in Fig. 5 through the uppermost point of the bar marked "Vacuum" will be the zero point for gas porosity for all of the gas melts. The bar "Vacuum" has two uppermost points, one shaded and the other unshaded, which were determined by using maximum density values obtained by different methods mentioned above. If a line is drawn through either point, comparisons must be made with corresponding points on bars representing soundness of the various gases. If the dotted line AA^1 is considered as the zero gas porosity line, the effect of the various gases on porosity is more noticeable. Carbon monoxide seems to have no effect and can be considered as either insoluble or soluble from room temperature to 1100°C. (2012°F.). The former is probably correct. Oxygen, nitrogen, carbon dioxide, sulphur dioxide, and air show gradual increases in porosity due to gas. The per cent voids of specimens melted under water vapor and hydrogen is more than twice that of the vacuum melted specimen, indicating a marked influence of these

Table 2

CHEMICAL ANALYSIS OF ALLOYS MELTED UNDER GAS ATMOSPHERES

(See Fig. 5)

Gas Atmosphere	*Heat Number*	*Cu., Per Cent*	*Sn., Per Cent*	*Zn., Per Cent*	*Ni., Per Cent*	*S., Per Cent*
Vacuum	9	87.15	9.15	2.76	0.75	
CO	16	87.35	9.16	2.63	0.75	
O_2	18	87.43	9.00	2.59	0.75	
N_2	19	87.41	8.96	2.79	0.75	
CO_2	13	87.09	9.03	2.98	0.75	
SO_2	14	87.09	8.96	2.80	0.75	0.10
Air	15	86.89	9.27	2.83	0.74	
H_2O	17	87.23	8.89	2.78	0.75	
H_2	12	87.36	9.14	2.70	0.75	
Leaky Castings	20	87.23	8.81	3.66		
	21	87.78	8.95	2.84		

two gases on porosity. Water vapor probably breaks down to form oxygen and free hydrogen, and it is quite likely that hydrogen alone is dissolved in the molten metal and precipitates during solidification to cause porosity.

36. Table 2 shows the chemical composition and heat numbers of the specimens charted in Fig. 5. Figures 6 and 7 show cross-sectional photographs of the specimens charted in Fig. 5. The effect of hydrogen and water vapor on porosity is obvious. A number of small cavities are evident in the specimens melted under air and sulphur dioxide. There is no apparent effect of carbon monoxide, oxygen, nitrogen, and carbon dioxide on porosity. The chemical analysis of the specimen melted under sulphur dioxide shows 0.10 per cent sulphur which was picked up during the melting process. Examination under the microscope revealed the presence of sulphides which seemed to have no bearing on porosity.

37. Before leaving the discussion of gas porosity, a short discussion of the practical application of these findings to foundry practices is in order. Virgin metals and scrap used in making the alloys are possible sources of gases, which may be carried over from previous refining or melting operations. Furthermore, they may come from moisture, oil, or corrosion products on the surface of the materials. Some or all of these, when heated, may ultimately break down to water vapor or hydrogen or both, and become absorbed by the molten metal. Furnace atmospheres in coal or coke fired natural draught pit type furnaces, oil, gas, or induction fur-

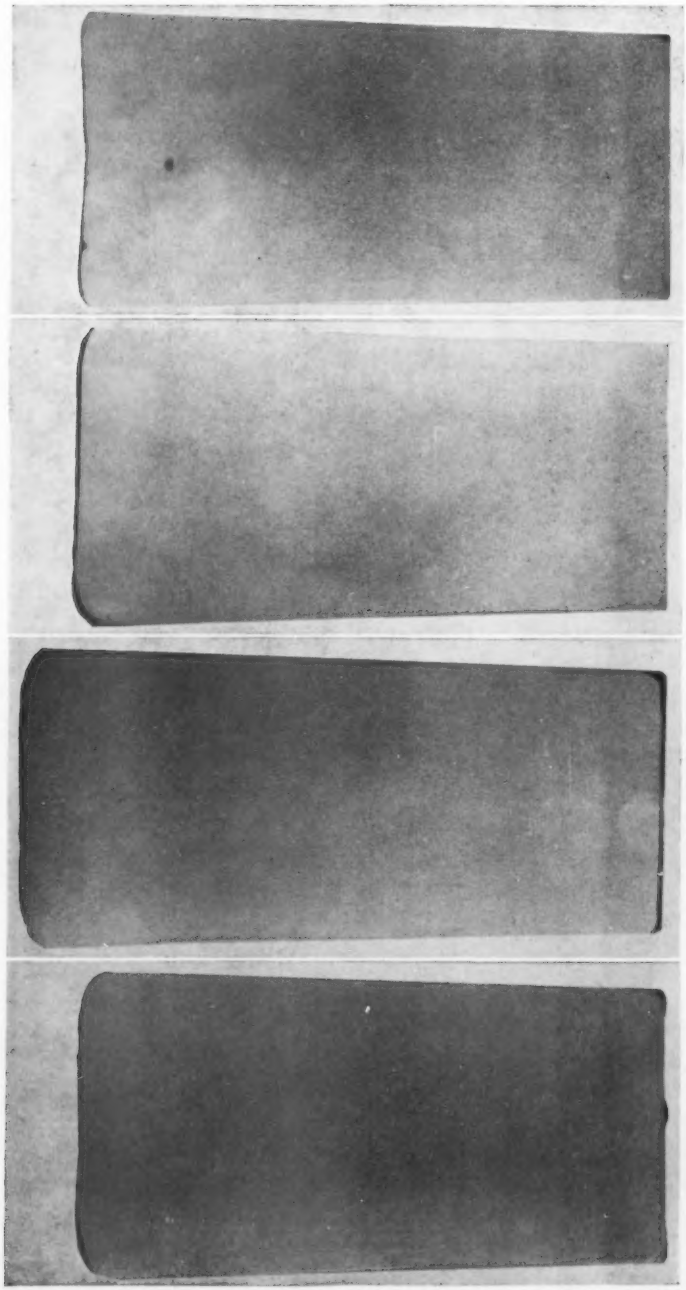
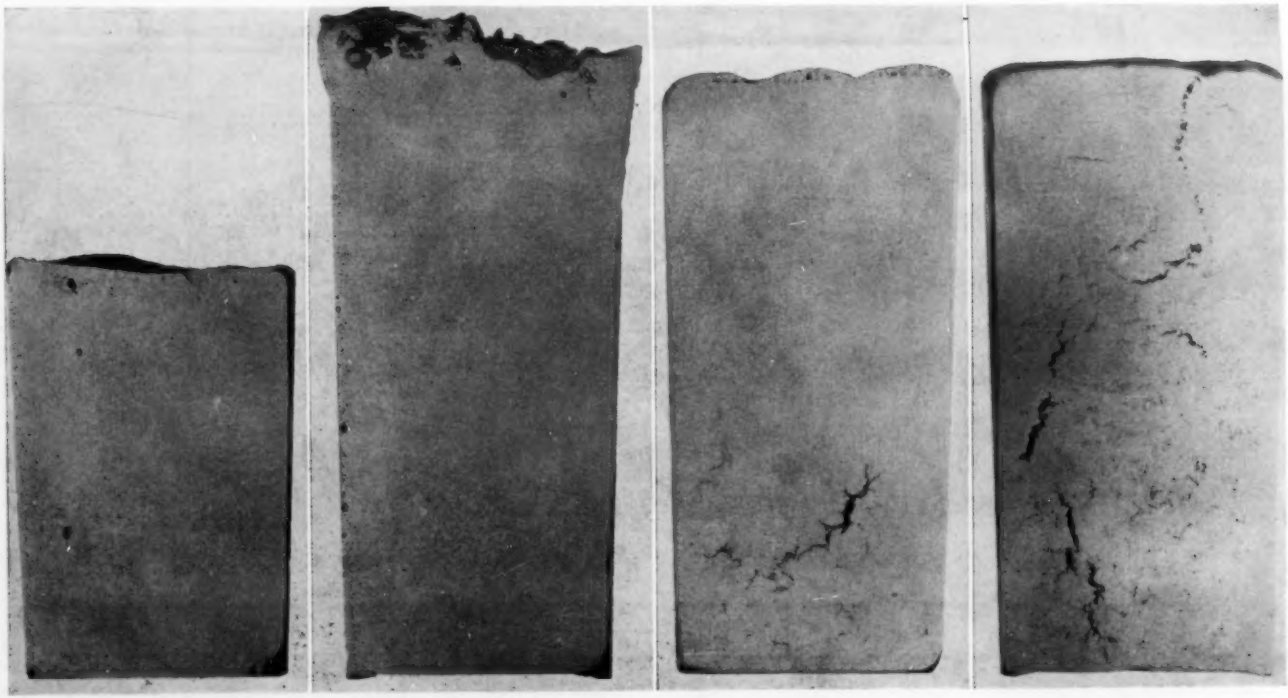


FIG. 6—LONGITUDINAL SECTIONS OF SPECIMENS OF TIN BRONZE MELTED AND SOLIDIFIED UNDER VARIOUS GAS ATMOSPHERES (SLIGHTLY REDUCED IN REPRODUCTION).



SULPHUR DIOXIDE AIR WATER VAPOR HYDROGEN

FIG. 7—LONGITUDINAL SECTIONS OF SPECIMENS OF TIN BRONZE MELTED AND SOLIDIFIED UNDER VARIOUS GAS ATMOSPHERES (SLIGHTLY REDUCED IN REPRODUCTION).

naces, may have hydrogen or water vapor present. Bailey[19] presented a chart showing gas atmospheres for a number of different types of copper melting furnaces, including some of those mentioned above, which showed that, in most cases, the hydrogen content varied between 0.0 to 2.0 per cent, but the water vapor was much higher, in one case having as much as 20.7 per cent. If the fluxes, deoxidizers or charcoal contain moisture, they also may be sources of gas porosity.

Study of Intercrystalline Shrinkage Porosity

38. *Effect of Alloying Elements.* Figure 8 shows the per cent voids and per cent soundness of a series of five alloys melted, solidi-

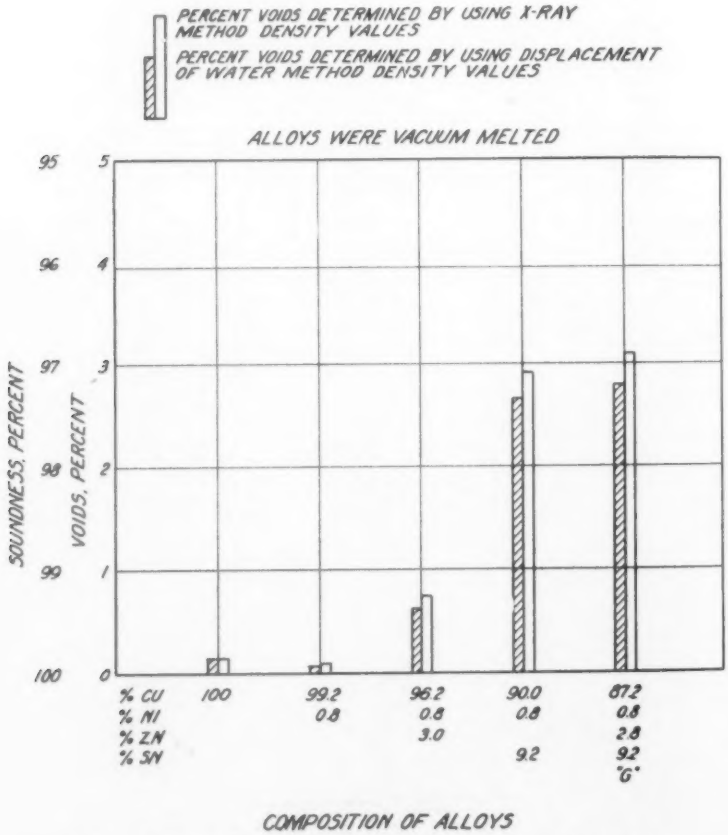


FIG. 8—EFFECT OF ELEMENTS USED IN TIN BRONZE ON SOUNDNESS.

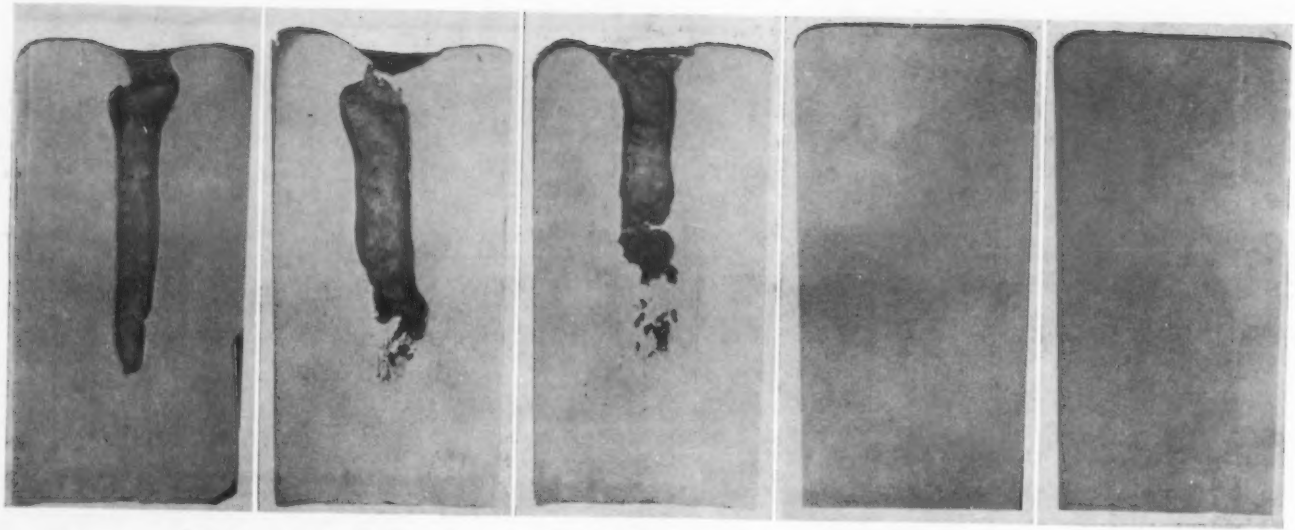
fied, and cooled in a graphite crucible in vacuo. The relatively low percentage of voids found in the first three alloys from the left side of the chart is worthy of note. These alloys do not contain tin. The percentage of voids found in the fourth and fifth alloys from the left side of the chart ranges from 4 to 30 times as much as the first three alloys. It is to be noted that the fourth and fifth alloys contain tin and that the fifth is tin bronze.

39. Cross-sectional photographs of these alloys are shown in Fig. 9. The alloys having low percentage of voids and no tin, show a pipe in the center of the specimen, while the alloys containing tin and having a high percentage of voids show no pipe whatsoever. It is obvious that the shrinkage in the alloys showing the pipe is external as evidenced by the pipe, whereas in the alloys containing tin, no external shrinkage is evident. Consequently, since some shrinkage has to take place, it can be assumed to be internal in the form of minute cavities. This is verified by the fact that the percentage of voids is relatively high and that minute cavities are discernible under the microscope.

40. In view of the results obtained from these data, it seems that, regardless of precise foundry control, it is impossible to manufacture castings entirely free of voids or porosity. It remains to be proved by hydraulic testing whether or not it is necessary for castings to be free of porosity if they are to be pressure-tight. However, it must be borne in mind that this factor is a natural phenomenon which cannot be easily controlled and presents a considerable handicap in the manufacture of pressure-tight castings.

41. *Influence of Pouring Temperatures.* Figure 10 shows the influence of pouring temperature on soundness for three compositions. Two curves are essentially tin bronze, one with and one without nickel. The third composition is a modification of tin bronze in which 6 per cent nickel is substituted for 6 per cent tin. Since the element tin has been shown to be a cause of intercrystalline or internal porosity in tin bronze, it was felt that, by substituting some other element for tin, internal porosity could be reduced. The substitution of nickel was made primarily to decrease intercrystalline shrinkage porosity and at the same time retain or improve all of the physical and corrosion resistant properties for which tin bronze is noted.

42. The tensile strength and elongation values shown are



	Per cent	Per cent	Per cent	Per cent	Per cent
COPPER	100	99.2	96.2	90.0	87.2
NICKEL		0.8	0.8	0.8	0.8
ZINC			3.0		2.8
TIN				9.2	9.2

FIG. 9—CROSS-SECTIONAL VIEW OF SPECIMENS USED FOR PREPARATION OF FIG. 8.

generally lower than what is normally expected for these alloys. A marked improvement in physical properties is obtained by using a standard test bar casting* rather than the test bar casting, Figure 2 (Top), used in this work. However, the results of these tests can be compared because possible variables were controlled for all melts. The important observation is not the actual values obtained but it is the trends taken by these alloys with variations in pouring temperature.

43. From the two curves in Fig. 10 representing tin bronze, two important observations can be made: (1) soundness increases

* Figure 10A, Navy Department Specification 46-B-28.

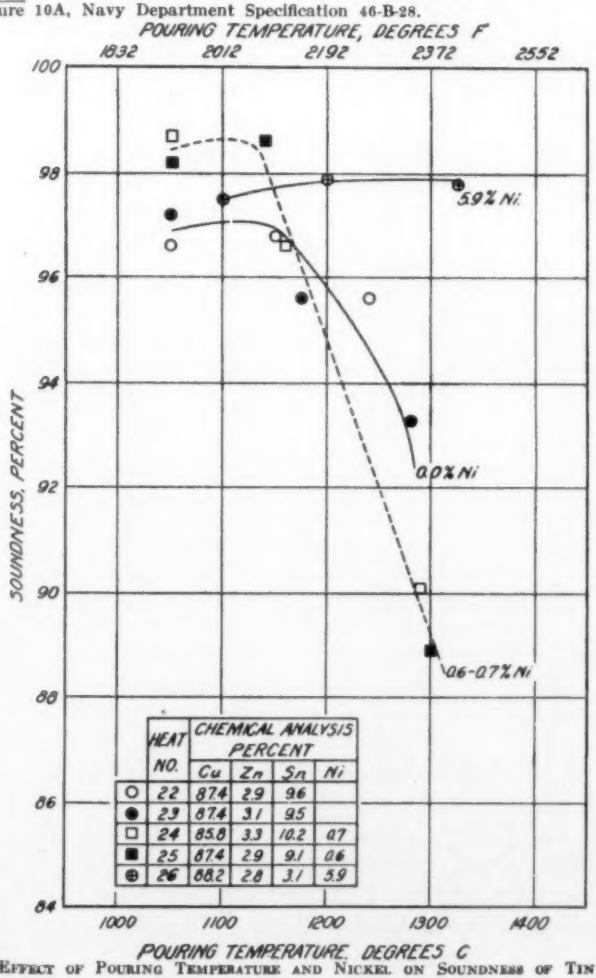


FIG. 10—EFFECT OF POURING TEMPERATURE AND NICKEL ON SOUNDNESS OF TIN BRONZE.

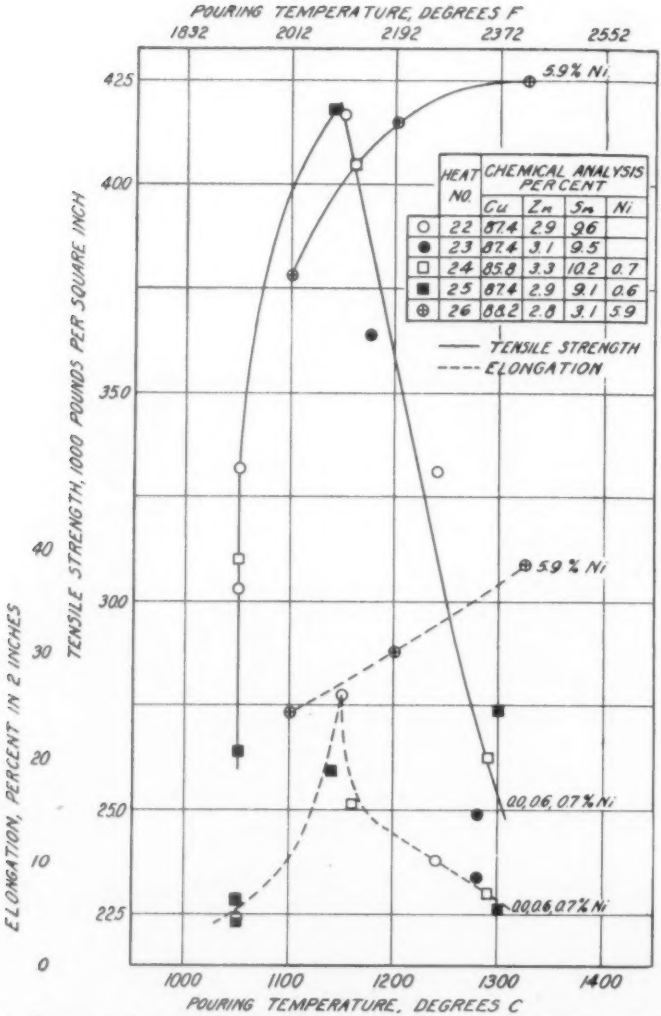


FIG. 11—EFFECT OF POURING TEMPERATURE AND NICKEL ON THE PHYSICAL PROPERTIES OF TIN BRONZE.

rapidly as the temperature decreases to 1150° C. (2102°F.) and is constant at lower temperatures; (2) the presence of 0.6 to 0.7 per cent nickel shows higher soundness at 1150°C. (2102°F.) and under, although the pouring temperatures above 1150°C. (2102°F.) seem to be more critical. Pouring temperature has little influence on soundness of the modified tin bronze alloy containing 6 per cent nickel.

44. In Fig. 11, pouring temperature is plotted against physical properties. Again the effect of pouring temperatures is to be noted on tin bronze. As the temperatures drop to 1150°C. (2102°F.), the physical properties increase sharply, and as the temperature drops below 1150°C. (2102°F.), the physical properties decrease. The physical properties of the modified tin bronze decrease with pouring temperature, but the change is not sharp, nor is there a critical pouring temperature.

45. Figures 10 and 11 are combined in Fig. 12, with soundness plotted against tensile strength and the pouring temperatures inserted at each point on the curves. The curves representing tin bronze show maximum tensile strength and soundness at a critical pouring temperature. Tin bronze containing from 0.6 to 0.7 per cent nickel shows improvement in soundness over tin bronze containing zero per cent nickel with no apparent change in tensile strength. The modified tin bronze alloy shows relatively little change in soundness with pouring temperature, although some loss in tensile strength is to be noted with lower pouring temperatures. However, the tensile strength values on the whole are equiv-

Table 3

EFFECT OF POURING TEMPERATURE ON VARIOUS PHYSICAL PROPERTIES

Specimen No.	*Heat Number*	*Pouring Temperature, °C*	*Pouring Temperature, °F*	*Average Density, gms/c.c. at 20°C**	*Per Cent Soundness**	*Tensile Strength, lb./sq. in.**	*Elongation, Per cent in 2 in.**
1	22	1240	2264	8.536	95.6	33,100	10.2
2	22	1150	2102	8.639	96.8	41,700	25.8
3	22	1050	1922	8.623	96.6	30,300	4.7
4	23	1280	2336	8.331	93.3	24,900	8.6
5	23	1175	2147	8.531	95.6	36,400	...
6	23	1050	1922	8.674	97.2	33,200	6.3
7	24	1290	2354	8.014	90.1	26,300	7.0
8	24	1160	2120	8.597	96.6	40,500	15.6
9	24	1050	1922	8.780	98.7	31,000	6.3
10	25	1300	2372	7.928	88.9	27,400	5.5
11	25	1140	2084	8.795	98.6	41,800	18.8
12	25	1050	1922	8.762	98.2	26,400	4.3
13	26	1325	2417	8.728	97.8	42,500	38.6
14	26	1200	2192	8.736	97.9	41,500	30.1
15	26	1100	2012	8.704	97.5	37,800	24.2

NOTE: * represents average of two specimens tested.

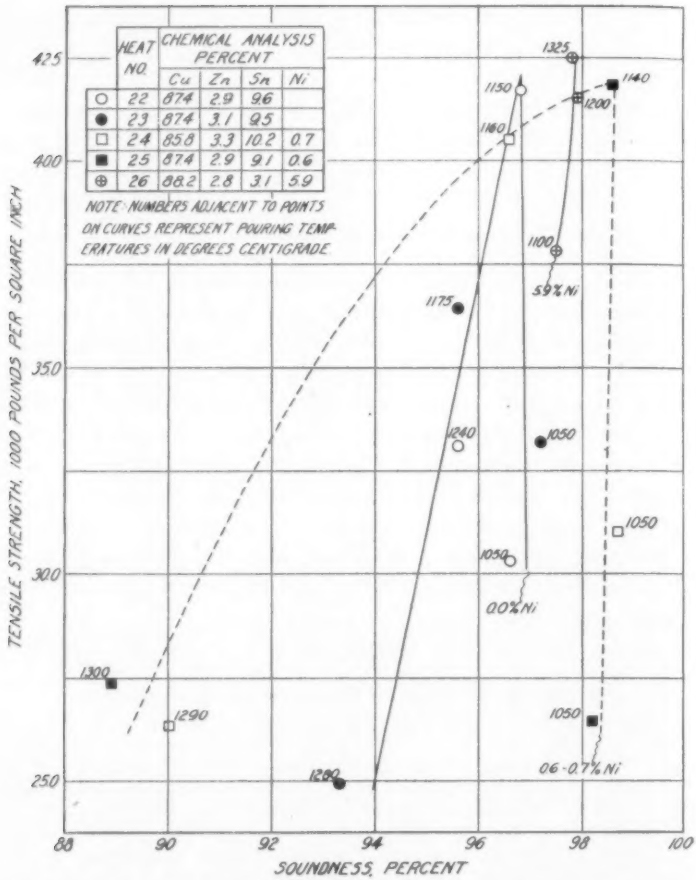


FIG. 12—RELATIONSHIP BETWEEN TENSILE STRENGTH, SOUNDNESS AND POURING TEMPERATURE AND THE INFLUENCE OF NICKEL ON THESE FACTORS IN TIN BRONZE.

alent to or higher than the maximum tensile strength values obtained on tin bronze. Table 3 shows the data used for preparing Figs. 10, 11 and 12. In addition, the average density of each specimen is included.

46. From the observation made on the curves (Figs. 10, 11 and 12), the importance of accurately controlling pouring temperatures of tin bronze in the foundry can be appreciated. To obtain maximum soundness and maximum physical properties, a casting has to be poured within a very narrow temperature range. The modified tin bronze, containing 6 per cent nickel, does not seem

to be affected by pouring temperatures. This may be due to the low tin content of the alloy. An alloy of this type would be of considerable help to the foundry, since it would eliminate one of the important factors in the manufacture of pressure-tight castings by removing the critical relationship between porosity and pouring temperature.

Hydraulic Pressure Testing

47. Table 4 shows the data obtained from the hydraulic pressure tests. It is to be noted that only one specimen, 10A, satisfactorily withstood pressure. The remainder leaked at 100 lb. pressure or less. Some of the other specimens had a higher soundness value

Table 4

HYDRAULIC PRESSURE TESTING DATA ON HOLLOW CYLINDER SPECIMENS

(Tin Bronze)

Chemical Analysis:

	Copper, Per Cent	*Tin, Per Cent*	*Zinc, Per Cent*
Heat No. 20.....	87.2	8.8	3.7
Heat No. 21.....	87.8	9.0	2.8

Pouring Temperature, °C	*Pouring Temperature, °F*	*Specimen No.*	*As Cast*		*Machined*		*Pressure Test**	
			Density at 20°C., gms/c.c.	*Per Cent Soundness*	*Density at 20°C., gms/c.c.*	*Per Cent Soundness*	*100 lb.*	*600 lb.*
1300	2372	1A	8.148	91.3	7.958	89.2	NG	
1300	2372	1B	8.170	91.6	8.080	90.6	NG	
1300	2372	6A	7.761	87.0				
1300	2372	6B	7.753	86.9	7.464	83.7	NG	
1210	2210	2A	8.309	93.1	8.129	91.1	NG	
1210	2210	2B	8.285	92.9	8.023	89.9	NG	
1200	2192	8A	8.493	95.2				
1200	2192	8B	8.465	94.9	8.340	93.5	NG	
1125	2057	10A	8.531	95.6	8.319	93.2	OK	OK
1125	2057	10B	8.534	95.7	8.346	93.5	NG	
1110	2030	3A	8.463	94.9	8.389	93.8	NG	
1110	2030	3B	8.471	95.0	8.284	92.9	NG	
1050	1922	4A	8.440	94.6	8.366	93.8	NG	
1050	1922	4B	8.428	94.5	8.428	94.3	NG	
1050	1922	12A	8.572	96.1	8.374	93.9	NG	
1050	1922	12B	8.549	95.8	8.343	93.5	NG	

* Only specimens withstanding 100 lb. per sq. in. pressure were tested at 600 lb. per sq. in. pressure.

NOTE: Specimens numbered 1A, 1B, 2A, 2B, 3A, 3B, 4A and 4B were taken from Heat No. 20, and specimens numbered 6A, 6B, 8A, 10A, 10B, 12A, and 12B were taken from Heat No. 21.

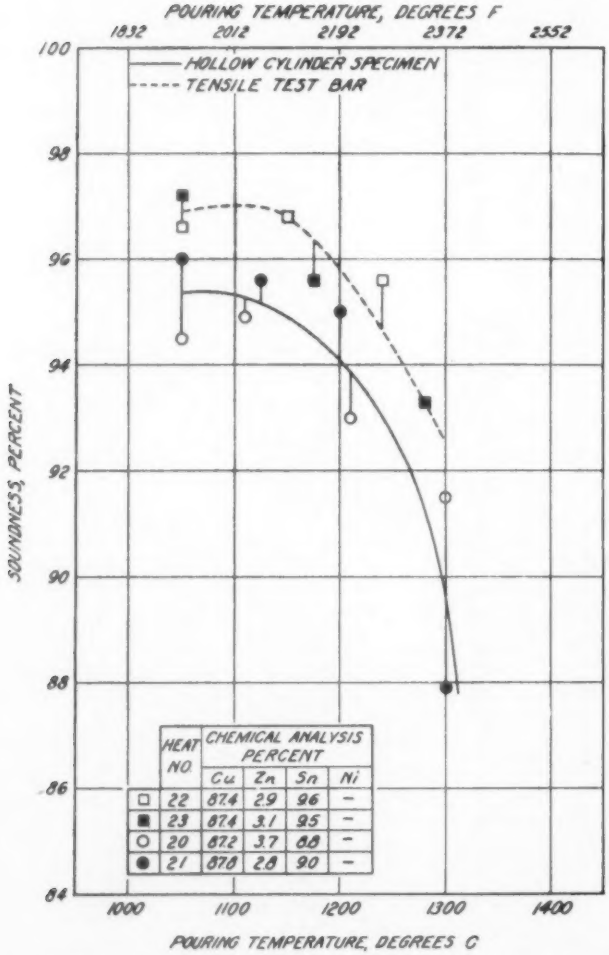


FIG. 13—EFFECT OF SHAPE AND DESIGN ON SOUNDNESS OF TIN BRONZE.

than 10A, but leaked when tested. Because of the deviation of results, no definite conclusions were drawn. Further work is contemplated on this type of testing with a view to the correlation of per cent soundness with pressure-tightness. It is felt that some changes in the design of the test casting will be necessary before satisfactory results can be obtained. Gardner and Saeger[20] conducted similar hydraulic pressure tests on cast red brass (85 per cent copper, 5 per cent zinc, 5 per cent tin, and 5 per cent lead)

with good results. Figure 13 shows the effect on soundness of two differently shaped and designed castings of identical composition.

SUMMARY AND CONCLUSIONS

48. Tin causes intercrystalline shrinkage porosity in tin bronze. This obviously suggests a change in composition or perhaps even a substitution of other alloying elements for tin.

49. Hydrogen and water vapor cause gas porosity in tin bronze. Sulphur dioxide seems to have a minor deleterious effect, while carbon dioxide, nitrogen, oxygen, and carbon monoxide cause little, if any, gas porosity.

50. Improper pouring temperatures cause porosity in tin bronze. Maximum soundness and physical properties can only be obtained within a very narrow pouring temperature range.

51. Modified tin bronze, in which 6 per cent nickel is substituted for 6 per cent tin, decreases the effect of pouring temperature on soundness and physical properties. In other words, high soundness values and good physical properties can be obtained over a wide pouring range, which should aid considerably in the production of pressure castings.

52. Design and shape of casting influence soundness.

ACKNOWLEDGMENTS

53. The author wishes to express appreciation to both Dr. R. H. Canfield and Dr. F. M. Walters for guidance in the work and for their many helpful criticisms and suggestions. Acknowledgment is also due Mr. James L. Darby and Mr. Thomas Cunningham for their assistance in casting and molding manipulations. The author wishes to express his thanks to the Navy Department for making the work possible and for permission to publish the results.

Bibliography

1—Primrose, H. S., *"Metallography as an Aid to the Brassfounder,"* JOURNAL, Institute of Metals, vol. 4, pp. 248-256 (1910).

2—Karr, C. P. and Rawdon, H. S. *"Standard Test Specimens of Zinc Bronze (88-10-2),"* U. S. Bureau of Standards, TECHNICAL PAPER No. 59, vol. 6 (1916).

3—Carpenter, H. C. H. and Elam, C. F., *"An Investigation of Unsound Castings of Admiralty Bronze (88-10-2): Its Cause and the Remedy,"* JOURNAL, Institute of Metals, vol. 19, pp. 155-175 (1918).

4—Smalley, O., *"Influence of Cored Structure,"* JOURNAL, Society of Chemical Industry, vol. 37, p. 191 (1918).

5—Bolton, J. W. and Weigand, S. A., *"Incipient Shrinkage in Some Non-Ferrous Alloys,"* TRANSACTIONS, A.I.M.E., Institute of Metals Div., vol. 83, pp. 475-491 (1929).

6—Daniels, E. J., *"Unsoundness in Bronze Castings,"* JOURNAL, Institute of Metals, vol. 43, pp. 125-142 (1930).

7—Dews, H. C., *"The Metallurgy of Bronze,"* London, Sir Isaac Pitman and Sons, Ltd. (1930).

8—Dews, H. C., *"The Effect of Phosphorus on the Strength of Admiralty Gun Metal,"* JOURNAL, Institute of Metals, vol. 44, pp. 255-266 (1930).

9—Crown, J. E., *"Bronze Pressure Castings,"* TRANSACTIONS, A.F.A., vol. 39, pp. 496-503 (1931).

10—Pilling, N. B. and Kihlgren, T. E., *"Some Effects of Nickel on Bronze Foundry Mixtures,"* TRANSACTIONS, A.F.A., vol. 39, pp. 93-113 (1931).

11—Doughty, E., *"Oxidation, A Cause of Porosity in Leaded Bronze,"* METALS AND ALLOYS, vol. 2, pp. 181-183, October (1931).

12—Bailey, G. L., *"Deoxidizers and Fluxes,"* THE METAL INDUSTRY (LONDON), vol. 40, No. 2, pp. 31-34, and No. 4, pp. 123-125 (1932).

13—Pilling, N. B. and Kihlgren, T. E., *"Casting Properties of Nickel Bronzes,"* TRANSACTIONS, A.F.A., vol. 40, pp. 289-305 (1932).

14—Fawcett, L. H., *"Properties of Bronze (88-10-2 and 88-8-4),"* NATIONAL METALS HANDBOOK, A.S.S.T., Cleveland, p. 1211 (1933).

15—Thews, E. R. *"Pouring Temperatures of Bronzes,"* THE FOUNDRY, vol. 62, March, p. 24; April, p. 24 (1934).

16—Allen, N. P. and Puddephat, S. H., *"Observations of the Porosity and Segregation of Two Bronze Ingots,"* JOURNAL, Institute of Metals, vol. 27, pp. 79-86 (1935).

17—Claus, W. and Bauer, R. W., *"Inverse Segregation and Gas Solubility in Tin Bronzes,"* METALLWIRTSCHAFT, vol. 15, pp. 587-600, June 26 (1936).

18—Jenkins, F. G., *"Zinc Bronze—Composition 'G',"* METALS AND ALLOYS, vol. 7, pp. 279-286 (1936).

19—Bailey, G. L., *"Effect of Melting Conditions on Gas Unsoundness in Metals,"* FOUNDRY TRADE JOURNAL, vol. 60, June 29, pp. 576-578, and July 6, pp. 9-10 (1939).

20—Gardner, H. B. and Saeger, Jr., C. M., *"The Effects of Antimony and Aluminum on Certain Properties of Cast Red Brass,"* TRANSACTIONS, A.F.A., pp. 423-443, December (1939).

Heat Treatment of Cast Iron

By Charles A. Nagler* and Dr. Ralph L. Dowdell*
Minneapolis, Minn.

Abstract

In this paper, the authors have investigated variations in hardness caused by the heat treatment of cast iron and have attempted to correlate the resulting microstructures with the hardness values obtained. Two types of heat treatment were used, namely, quenching and tempering and isothermal transformation by quenching in a salt bath at elevated temperatures. Two irons were used: (1) a base iron of the analysis given in Table 1, and (2) the same iron to which 0.98 per cent molybdenum had been added. These irons were cast into 1.2-in. diameter test bars which were machined to 1.0-in. diameter and then cut to 1.0-in. lengths. The test bars were polished to secure a smooth surface and a high rate of thermal conductivity. The isothermally transformed samples were heated to 1600°F. and quenched in a salt bath at the desired transformation temperature. The quenched and tempered samples were held at the same temperature and quenched into water at 60°F. They were then dried and placed in a salt bath for tempering at the proper temperature and then quenched in water at 60°F. Both the quench and temper treatments and the isothermal transformation treatment were carried out over a wide range of temperatures. The results of the investigation are given in the paper. The authors conclude that the hardness curves for both irons when isothermally transformed are parallel over a temperature range of 200-1300°F, that the same holds for the two irons over the same range when subjected to the quenching and tempering treatment, that the addition of molybdenum delays the rate of austenite transformation and increases the hardness of the samples after one hour of isothermal transformation, that the hardness of both the base and alloyed irons is higher after the quenching and

* University of Minnesota, Department of Metallography.
Note: This paper was presented before a Gray Cast Iron session at the 45th Annual A.F.A. Convention, New York City, N. Y., May 15, 1941.

tempering treatment than after the isothermal transformation for the same period of time and the same temperature.

1. Heat treatment is used to improve the physical properties of cast iron. The purpose of this investigation is to study the variations in hardness caused by heat treatment, and to correlate resulting microstructures with hardness values. The two types of heat treatment used in this investigation were the quench and temper, and the isothermal transformation by quenching into a salt bath at an elevated temperature.

2. Cast iron used in this investigation was poured from a commercial heat of electric furnace iron. The bars were cast from a 200 lb. heat, and 50 lb. ladles were used. A late addition of 0.75 per cent silicon was added to each hand ladle (iron melted with 1.25 per cent silicon) to arrive at the desired silicon content. In the case of the molybdenum iron bars, ferromolybdenum and ferrosilicon were added simultaneously to the hand ladle and the molten metal from the furnace was poured on top of them. The test bars used in this investigation were poured according to A.S.T.M. standards. The analyses of the two irons used in this investigation are given in Table 1.

3. The 1.2-in. diameter companion test bars were machined to 1.00-in. diameter and polished with emery cloth to insure a smooth surface with a high rate of thermal conductivity. Before machining the bars, the portion adjacent to the pouring basin was marked, and will be referred to as the top of the test bar. One-half-in. was removed from the bottom of each bar before sampling. Samples were cut from the bars in one-in. lengths. The relative position of samples taken from the transverse bars is shown in Table 2.

Table 1

ANALYSES OF IRONS USED

	Base Iron *Iron No. 1*	*Base Iron plus Molybdenum* *Iron No. 2*
Total carbon, per cent	3.06	2.97
Combined carbon, per cent	0.79	0.86
Graphite, per cent	2.27	2.11
Manganese, per cent	0.67	0.65
Silicon, per cent	2.28	2.26
Molybdenum, per cent	0.00	0.98
Phosphorus, per cent	0.139	0.134
Sulphur, per cent	0.083	0.082

Table 2

POSITION OF SAMPLES

Distance from Bottom, in.	*Base Iron*		*Base Iron plus Molybdenum*	
	Sample	*Sample*	*Sample*	*Sample*
1	1 AQ	1 AQ D	2 AQ	2 AQ D
2	1 200	1 200 D	2 200	2 200 D
3	1 300	1 300 D	2 300	2 300 D
4	1 400	1 400 D	2 400	2 400 D
5	1 500	1 500 D	2 500	2 500 D
6	1 600	1 600 D	2 600	2 600 D
7	1 700	1 700 D	2 700	2 700 D
8	1 800	1 800 D	2 800	2 800 D
9	1 900	1 900 D	2 900	2 900 D
10	1 1000	1 1000 D	2 1000	2 1000 D
11	1 1100	1 1100 D	2 1100	2 1100 D
12	1 1200	1 1200 D	2 1200	2 1200 D
13	1 1300	1 1300 D	2 1300	2 1300 D
14	1	1 D	2	2 D

Preparation of Samples

4. Samples for heat treatment were prepared in the following manner. The samples were surface ground to a length of 1.00-in., and a hole drilled in the center of the sample with a 3/64-in. drill to a depth of 7/16-in. One-quarter-inch from the center along a diameter, a hole of ⅛-in. diameter was drilled to a depth of ¼-in., for a 1-in. length of low carbon steel, which was used as a sample holder and on which the thermocouple was secured.

5. Figure 1 shows the completely assembled sample. The thermocouple was luted into the hole with asbestos wool. The top

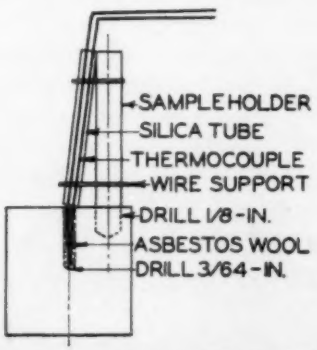


FIG. 1—COMPLETELY ASSEMBLED SAMPLE SHOWING SAMPLE HOLDER AND THERMOCOUPLE IN POSITION.

of the sample was covered with alundum cement to prevent any of the graphite, in which the sample was packed, from attacking the thermocouple.

Heat Treatment Procedure

6. Heat treatment of the samples was carried out in the following manner. The muffle furnace was heated to 1600°F. and held at temperature by an automatic controller. The sample to be heat-treated was packed in graphite in a small high alloy steel box.

7. Samples 1 AQ, 1 AQ D, 2 AQ, 2 AQ D, were heat treated in the following manner. Heat to 1600°F. in one hour, hold at temperature for one hour, and then rapidly quench into water at 60°F. with a rapid stirring action.

8. The isothermally-transformed samples were given the following treatment, and will be designated without any letter following the number (1 200 sample isothermally transformed at 200°F.). The samples were heated in a high alloy steel box with the inside dimensions of 1¾ x 3¾ x 1½-in. One sample was heat-treated at a time. The temperatures of the samples were checked at frequent intervals during the period that they were held at temperature. The samples were packed in graphite to prevent decarburization during heat treatment. After the samples had been at 1600°F. for one hour, they were removed rapidly from the muffle furnace and quenched into a salt bath at the desired transformation temperature with a quick stirring action. The samples were allowed to remain in the salt bath for one hour, and at the end of that period, were removed and quenched into water at 60°F.

9. The quenched and tempered samples were given the following treatment, and will be referred to with the letter D on each sample (1 200 D sample quenched and tempered at 200°F). The samples were heated, and held at temperature by the same method as described for the isothermally-transformed samples. After the samples had been at temperature for one hour, they were removed rapidly from the muffle furnace and quenched into water at 60°F. The samples were dried and put into a salt bath for tempering. They were then stirred around in the salt bath when charged for tempering, held in the tempering bath for one hour, and then quenched into water at 60°F.

10. The isothermal transformation and the quench and temper heat treatments were carried out over a temperature range of

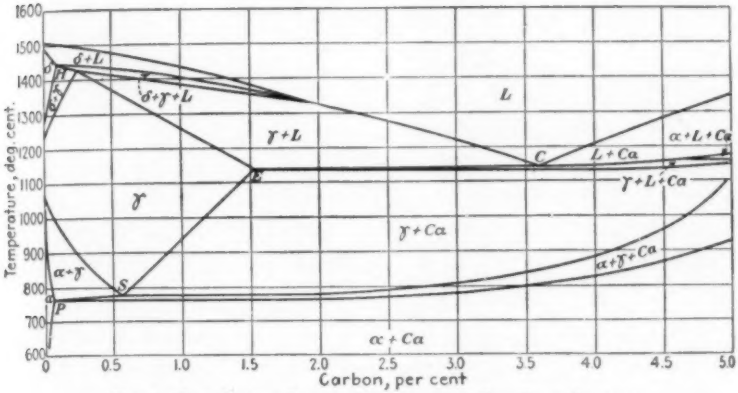


FIG. 2—IRON-CARBON-SILICON DIAGRAM AT 2 PER CENT SILICON.

200 to 1400°F. The low-temperature temperings and isothermal transformations of 200 and 300°F. were carried out in an oil bath. The higher-temperature tempering and isothermal transformations were carried out in a salt bath.

Thermal Analysis of the Irons Used

11. To definitely establish the critical points in the irons studied, thermal analysis was used. Thermal analysis indicated that the critical range for Iron No. 1 was in the range of 1344 to 1304°F., and for Iron No. 2 in the range of 1315 to 1304°F. The 1600°F. temperature for solution treatment was used in order that, on quenching from that temperature, martensite is formed as the matrix with the graphite flakes imbedded in it. Figure 2[1] illustrates a section of iron-carbon-silicon diagram at 2 per cent silicon.

Explanation of Structural Changes in Cast Iron

12. On examination of Fig. 2, it can be seen that at the temperature of 1400°F., where the eutectoid temperature occurs, there is not a direct transformation of the austenite to pearlite, but a region in which austenite, ferrite and cementite exist. The existence of this three-phase region is due to the addition of 2 per cent silicon to the normal Fe-Fe_3C system*.

13. On heating a sample of cast iron to the solution temperature of 1600°F., the following occurs. As the sample is heated from

[1] Greiner, E. S., Marsh, J. S., and Stoughton, B., *"The Alloys of Iron and Silicon,"* Alloys of Iron Research, Monograph Series, p. 140.

* Diagram for the normal system may be found in any modern treatise on metallurgy.

room temperature to 1400°F., the solubility of carbon in the ferrite increases from about 0.006 to 0.035 per cent. On further heating to 1500°F., the solubility of the carbon in the austenite increases to 0.70 per cent, and also the sample is heated through the region in which the austenite, ferrite and cementite co-exist. On further heating above 1500°F., all the ferrite dissolves in the austenite, and at the solution temperature, only austenite and cementite co-exist. Solubility of carbon in austenite at 1600°F. is about 0.80 per cent. Holding the sample at 1600°F. for one hour allows time for partial solubility of the carbides in the austenite.

14. It was thought advisable to hold the solution temperature constant at 1600°F. so that the micro-constituents and hardnesses could be compared after the samples had been given a variety of heat treatments. The amount of carbon dissolved in the austenite will influence the hardness obtained on quenching cast iron from above the critical temperature.

15. In the description thus far, no mention has been made of the graphitic carbon which is present in the iron. The major point of interest in this work is with the matrix of the iron rather than with the graphite and graphitic distribution. In all probability, graphite is one of the controlling factors in the hardening of the iron.

16. On heating a sample of cast iron to 1600°F., the graphite remains relatively unchanged, with a possibility of diffusion of some of the carbon into the austenite, to the maximum solubility at that temperature, from the graphite flakes after holding at the high temperature.

17. To make a complete study as to the change in hardness and microconstituents with heat treatment, samples were isothermally transformed, and quenched and tempered.

Hardness vs. Structure in As-Cast Condition

18. Examination of the base iron (Iron No. 1), in the as-cast condition, disclosed that the iron was a normal gray iron. Figure 3 (Upper Left) shows a typical section of the iron containing pearlite, graphite, cementite, and some inclusions.

19. Iron No. 2, which is Iron No. 1 with the molybdenum addition, showed the presence of ferrite along the graphite flakes. As shown in Fig. 3 (Upper Right), the pearlite was very finely

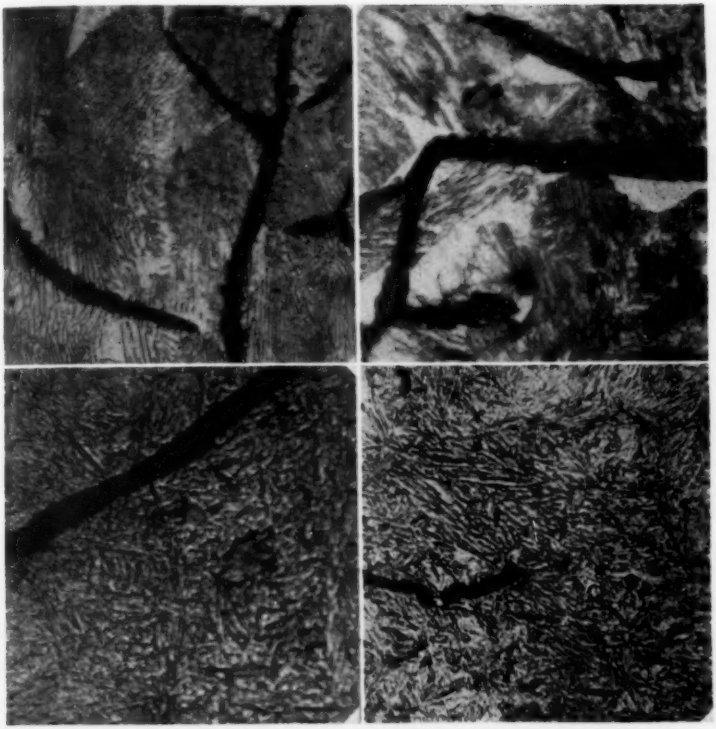


FIG. 3—(UPPER LEFT)—BASE IRON (IRON NO. 1) IN THE AS CAST CONDITION. FIRTH HARDNESS 198. x1000. (UPPER RIGHT)—ALLOYED IRON (IRON NO. 2) IN THE AS CAST CONDITION. FIRTH HARDNESS 250. x1000. (LOWER LEFT)—IRON NO. 1 WATER QUENCHED FROM 1600°F. FIRTH HARDNESS 529. x1000. (LOWER RIGHT)—IRON NO. 2 WATER QUENCHED FROM 1600°F. FIRTH HARDNESS 590. x1000.

laminated, and there appeared to be areas of excess ferrite in the matrix of the metal, associated with the graphite flakes. The presence of the ferrite around the graphite flake would tend to indicate that some graphitization had taken place in the iron while cooling from the eutectic temperature.

20. The Firth diamond hardness number of the Iron No. 2 was found to be 250, and that of Iron No. 1, 198. (The Firth hardness numbers and Brinell hardness numbers are in very close agreement and, due to the wide range in hardness values determined, it was deemed desirable to use Firth hardness numbers.) The hardness value in all cases is an average of 10 hardness readings taken across the diameter of the sample at a distance of 10th's of an inch apart. For the greater part of the hardnesses deter-

mined, across a sample, there was good agreement, even though soft graphite flakes were present in the matrix.

Preparation of Samples for Hardness Tests

21. The hardnesses of Irons Nos. 1 and 2 were determined in the as quenched conditions. The samples were heated to 1600°F. in one hour, and then held at temperature for one hour before quenching. On quenching, the samples were removed rapidly from the furnace and given a stirring action in the water bath. The samples were then sectioned into two equal parts by the use of a cut off wheel, surface ground to insure a level surface, polished through the rouge wheel, and then hardness determinations were made with the Firth Hardometer. In all cases, the load of 30 Kg. was applied for 30 sec. This method of sample preparation for hardness determination was repeated in all samples studied.

Structural Hardness of Iron No. 1

22. Iron No. 1, on quenching, was found to have a Firth Hardness value of 529, while Iron No. 2 was found to have a hardness value of 590. Photomicrographs of the hardened irons are shown in Fig. 3 (Lower Left and Right, respectively) and indicate the matrix of each iron to be fully martensitic, plus the graphite flakes. The martensite produced is not as typical as the martensite ordinarily found in steel. The slip planes show less distinct markings of the martensite making the angles of 60° and 120° less distinct. This martensite is typical of that found in a fine grained tool steel that has been quenched from just above the eutectoid temperature.

23. Figures 4 and 5 illustrate the changes in microstructure of the base iron (Iron No. 1) as the isothermal transformation temperature* is raised from 200 to 1300°F. The hardness of the base iron was 529 when quenched into water from 1600°F., and on isothermal transformation at 200°F., the hardness dropped to 505. On further transformation of the iron at progressively higher temperatures, the hardness dropped to 229 at 700°F. Figure 4 (Upper Left and Right) shows that the irons transformed at 200 and 300°F., respectively, for one hour have a martensitic matrix plus some retained austenite. The hardness of the 200°F. sample

*Base iron samples did not crack on quenching from 1600°F. to the isothermal transformation temperature.

was 505, while in the case of the 300°F. sample, the hardness was found to be 487. This would tend to indicate that the 300°F. sample had more of its carbides agglomerated than in the 200°F. sample, after holding for the same length of time at their respective transformation temperatures.

24. At 400°F., the product of transformation is the typical acicular troostite structure (Fig. 4—Center Left). The length of the martensitic needles in Fig. 4—Center Left appears much greater than is evident in either the quenched specimen (Fig. 3—Lower Left) or in irons transformed at the lower temperatures (Fig. 4—Upper Left and Right). Due to the poorly defined needles produced on etching the martensite in Fig. 4—Upper Left and Right it is difficult to determine the length of the needles and the size of the original austenitic grain.

25. Transformation of the base iron at 500°F. resulted in the disappearance of the fine black needles, and the beginning of the coagulation of the carbides. The carbide coagulation continues as the isothermal transformation temperature is increased up to 1000°F. It is at this temperature (1000°F.) that fine pearlite appears, as a transformation product (Fig. 5—Center Left).

26. The hardness curve (Fig. 8), shows that, over the range of 700 to 900°F., there is a decrease and then a subsequent increase in hardness. By referring to Fig. 4—Lower Right and Fig. 5—Upper Left and Right, it is rather difficult to explain this dip in the hardness curve. As the transformation temperature is increased from 1000 to 1300°F., the fine pearlite that was evident at 1000°F. (Fig. 5—Center Left) tends to form a diffused pearlite (spheroidized carbide plus ferrite) at 1300°F. (Fig. 5—Lower Right).

Structure vs. Hardness in Iron No. 2

27. In the case of the base iron to which molybdenum was added (Iron No. 2), the hardness on isothermal transformation* appears to have the same trend as that of Iron No. 1 (Fig. 8). At the transformation temperatures of 200 and 300°F., the micrographs show a matrix of martensite plus some retained austenite (Fig. 6—Upper Left and Right).

28. At 400°F., an acicular troostite structure appears (Fig. 6—Center Left) which is similar to the product formed at 400°F.

* Alloyed cast iron samples did not crack on quenching from 1600°F. to the isothermal transformation temperature.

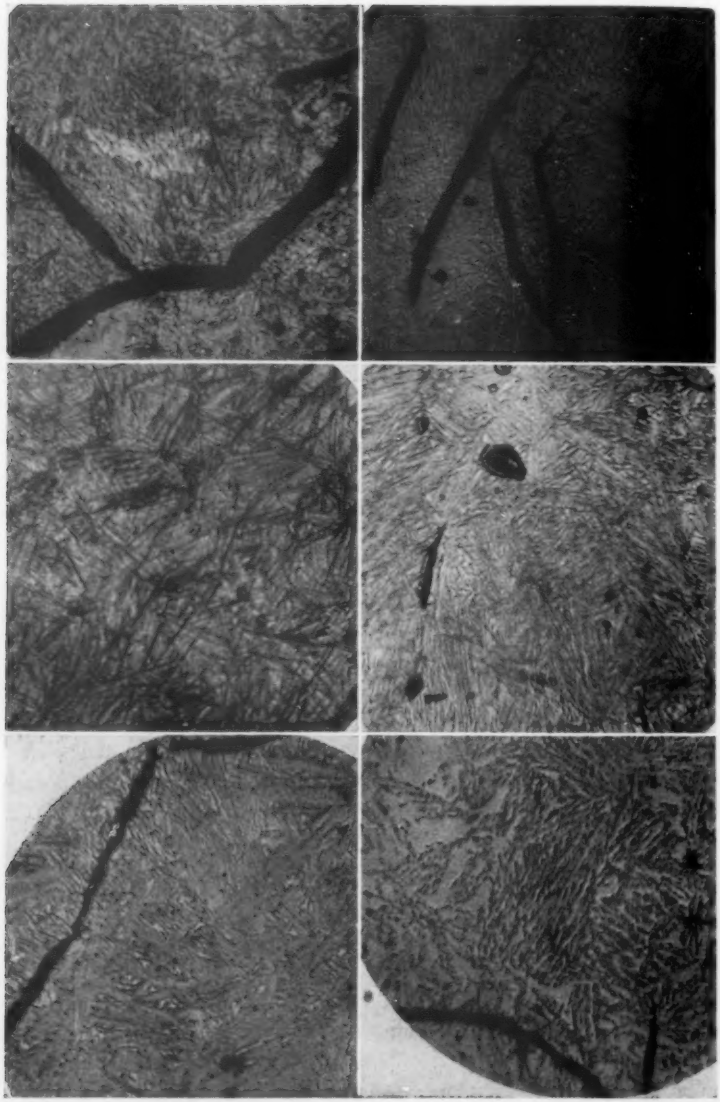


FIG. 4—BASE IRON (IRON NO. 1) ISOTHERMALLY TRANSFORMED FROM 1600°F. FOR ONE HOUR AT TEMPERATURES SHOWN. x1000. (UPPER LEFT)—TRANSFORMED AT 200°F. FIRTH HARDNESS 505. (UPPER RIGHT)—TRANSFORMED AT 300°F. FIRTH HARDNESS 487. (CENTER LEFT)—TRANSFORMED AT 400°F. FIRTH HARDNESS NUMBER 422. (CENTER RIGHT)—TRANSFORMED AT 500°F. FIRTH HARDNESS 347. (LOWER LEFT)—TRANSFORMED AT 600°F. FIRTH HARDNESS 305. (LOWER RIGHT)—TRANSFORMED AT 700°F. FIRTH HARDNESS 229.

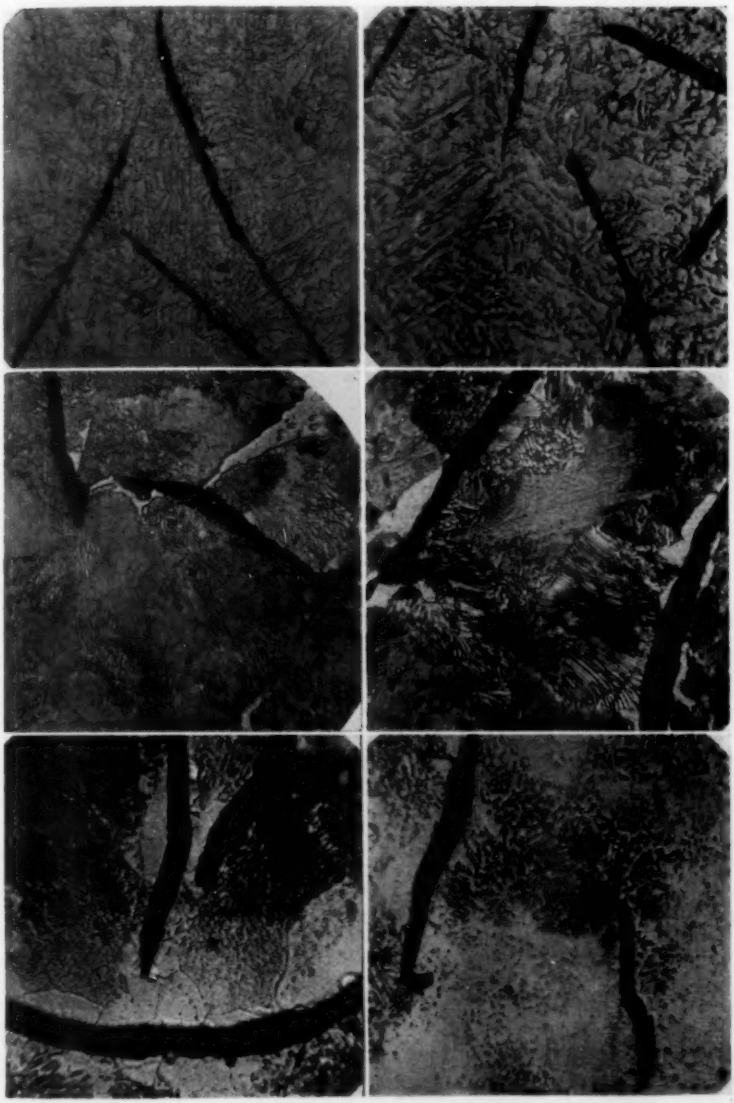


FIG. 5—BASE IRON (IRON NO. 1) ISOTHERMALLY TRANSFORMED FROM 1600°F. FOR ONE HOUR AT TEMPERATURES SHOWN. x1000. (UPPER LEFT)—TRANSFORMED AT 800°F. FIRTH HARDNESS 233. (UPPER RIGHT)—TRANSFORMED AT 900°F. FIRTH HARDNESS 255. (CENTER LEFT)—TRANSFORMED AT 1000°F. FIRTH HARDNESS 250. (CENTER RIGHT)—TRANSFORMED AT 1100°F. FIRTH HARDNESS 223. (LOWER LEFT)—TRANSFORMED AT 1200°F. FIRTH HARDNESS 164. (LOWER RIGHT)—TRANSFORMED AT 1300°F. FIRTH HARDNESS 143.

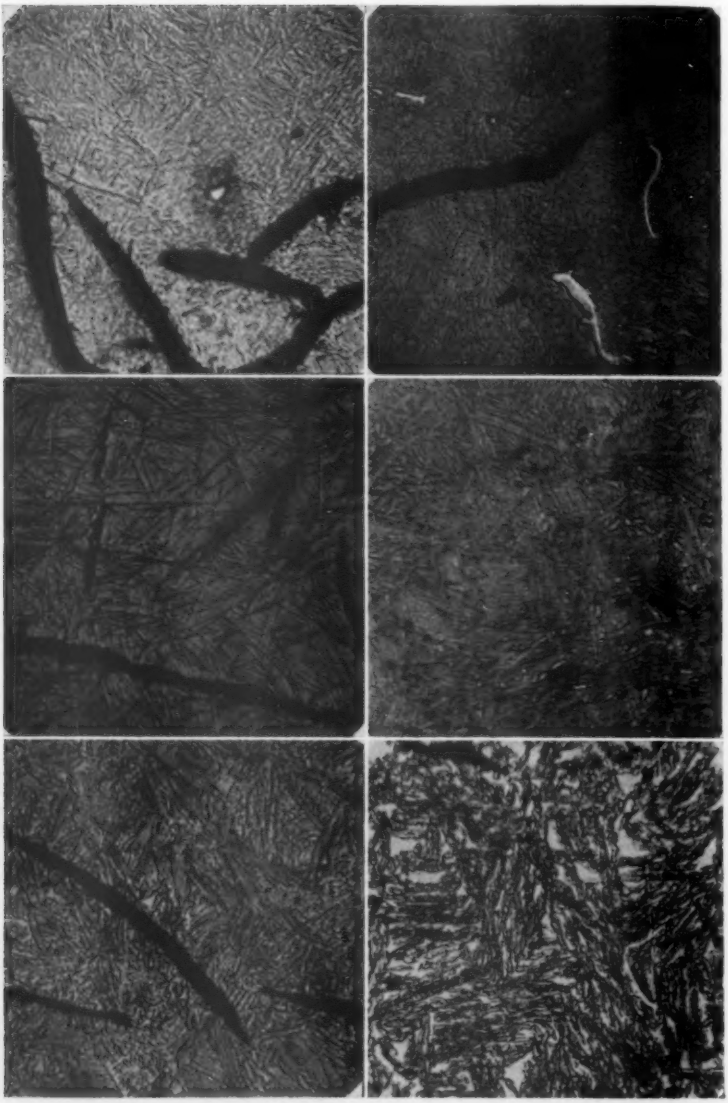


FIG. 6—ALLOYED IRON (IRON NO. 2) ISOTHERMALLY TRANSFORMED FROM 1600°F. FOR ONE HOUR AT TEMPERATURES SHOWN. X1000. (UPPER LEFT)—TRANSFORMED AT 200°F. FIRTH HARDNESS 550. (UPPER RIGHT)—TRANSFORMED AT 300°F. FIRTH HARDNESS 513. (CENTER LEFT)—TRANSFORMED AT 400°F. FIRTH HARDNESS 494. (CENTER RIGHT)—TRANSFORMED AT 500°F. FIRTH HARDNESS 432. (LOWER LEFT)—TRANSFORMED AT 600°F. FIRTH HARDNESS 340. (LOWER RIGHT)—TRANSFORMED AT 700°F. FIRTH HARDNESS 286.

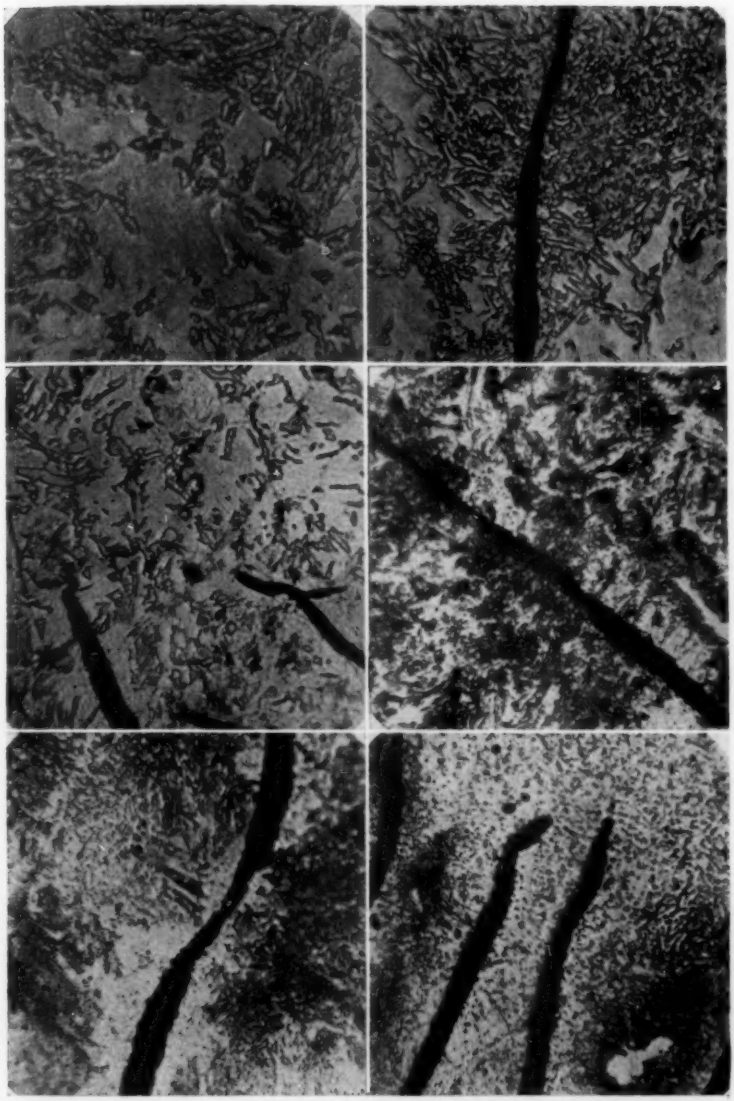


FIG. 7—ALLOYED IRON (IRON NO. 2) ISOTHERMALLY TRANSFORMED FROM 1600°F. FOR ONE HOUR AT TEMPERATURES SHOWN. X1000. (UPPER LEFT)—TRANSFORMED AT 800°F. FIRTH HARDNESS 263. (UPPER RIGHT)—TRANSFORMED AT 900°F. FIRTH HARDNESS 239. (CENTER LEFT)—TRANSFORMED AT 1000°F. FIRTH HARDNESS 288. (CENTER RIGHT)—TRANSFORMED AT 1100°F. FIRTH HARDNESS 274. (LOWER LEFT)—TRANSFORMED AT 1200°F. FIRTH HARDNESS 191. (LOWER RIGHT)—TRANSFORMED AT 1300°F. FIRTH HARDNESS 172.

in Iron No. 1 (Fig. 4—Center Left). The base iron (Iron No. 1) had a hardness of 422 at 400°F., while the alloy iron (Iron No. 2) had a hardness of 494.

29. As the transformation temperature of the alloyed iron is increased from 500 to 1000°F., the carbides coagulate. At 1100°F. (Fig. 7—Center Right), spheroidized carbides, form and as the temperature is increased to 1300°F., ferrite areas increase, as does carbide particle size.

HARDNESS VS. QUENCHING AND TEMPERING

30. The latter part of the investigation was devoted to a comparison of the hardness produced on a quenching and tempering heat treatment as given to the base iron (Iron No. 1) and the alloyed iron (Iron No. 2). The same solution temperature (1600°F.) and time of solution (one hour) were used. In this part of the investigation, all specimens were quenched to a fully martensitic structure and then tempered at successively higher temperatures up to 1300°F. A one hour tempering time at temperature was used on all specimens.

Results of Treatment on Iron No. 1

31. Figures 9 and 10 illustrate the structures formed on quenching and tempering of Iron No. 1. The photomicrographs

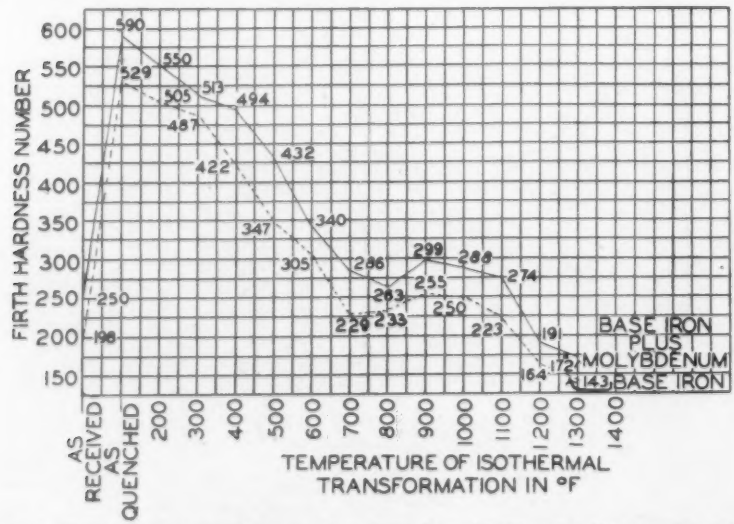


FIG. 8—HARDNESS OF BASE IRON AND BASE IRON PLUS MOLYBDENUM TRANSFORMED FROM 1600°F. FOR ONE HOUR AT TEMPERATURES SHOWN.

show that, as the tempering temperature is increased, the carbides coagulate and ferrite areas form.

32. At 1100°F. (Fig. 10—Center Right), there is evidence of the first appearance of a fully spheroidized carbide. As the temperature is increased to 1300°F., the carbide particles increase in size.

33. The quenched, fully-martensitic base iron had a hardness of 524, and this value dropped to 172, at a tempering temperature of 1300°F. The change in hardness of the base iron on a one hour tempering treatment at successively higher temperatures is shown in Fig. 13.

Results of Treatment on Iron No. 2

34. The alloy iron (Iron No. 2) also was given the quench and temper heat treatment. Figures 11 and 12 show the microstructures produced by increasing the tempering temperature of the quenched alloy iron from 200 to 1300°F. The comparison of the photomicrographs of the base iron and the alloyed iron show that the structures produced at the different tempering temperatures are very similar. By reference to Fig. 13, it is evident that the hardness curves of the two irons are parallel over the tempering range studied. In the case of the alloyed iron, there is very little change in hardness over the tempering range of 400 to 700°F. Fig. 11 Center Left to Lower Right show very little change in microstructure. Similarly, in the case of the base iron, the hardness remains unchanged over the range of 400 to 600°F., and no change is evident in the microstructure (Fig. 9—Center Left to Lower Left).

Mechanism of Isothermal Transformation

35. It would be of interest to analyze what takes place during isothermal transformation. The austenite, that is formed at the solution temperature, is quenched rapidly to some subcritical temperature and is forced to transform to its decomposition product at this temperature. To illustrate this mechanism, an isothermal transformation curve has been chosen which is theoretical in many respects, but which will fit in well with the explanation. This curve is shown in Fig. 14. If a cooling rate from above the eutectoid temperature similar to curve *A* is chosen, martensite will be formed upon quenching. If the cooling rate be represented by

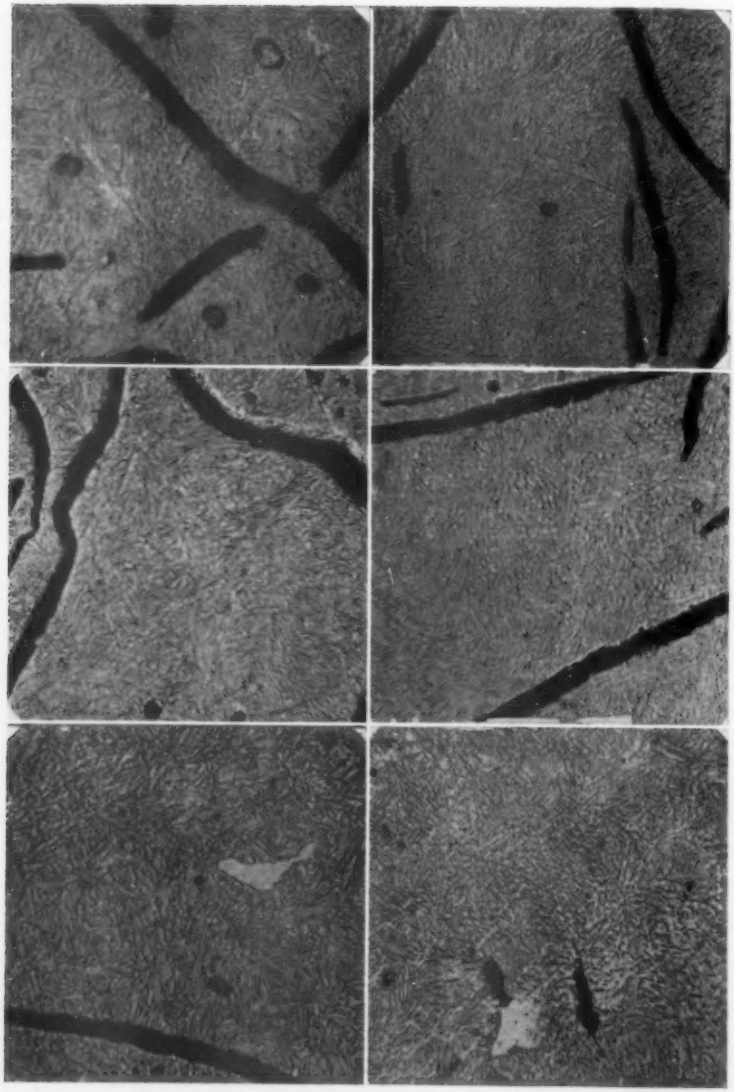


FIG. 9—BASE IRON (IRON NO. 1) QUENCHED FROM 1600°F. AND TEMPERED FOR ONE HOUR AT TEMPERATURES SHOWN. x1000. (UPPER LEFT)—TEMPERED AT 200°F. FIRTH HARDNESS 510. (UPPER RIGHT)—TEMPERED AT 300°F. FIRTH HARDNESS 488. (CENTER LEFT)—TEMPERED AT 400°F. FIRTH HARDNESS 451. (CENTER RIGHT)—TEMPERED AT 500°F. FIRTH HARDNESS 450. (LOWER LEFT)—TEMPERED AT 600°F. FIRTH HARDNESS 448. (LOWER RIGHT)—TEMPERED AT 700°F. FIRTH HARDNESS 422.

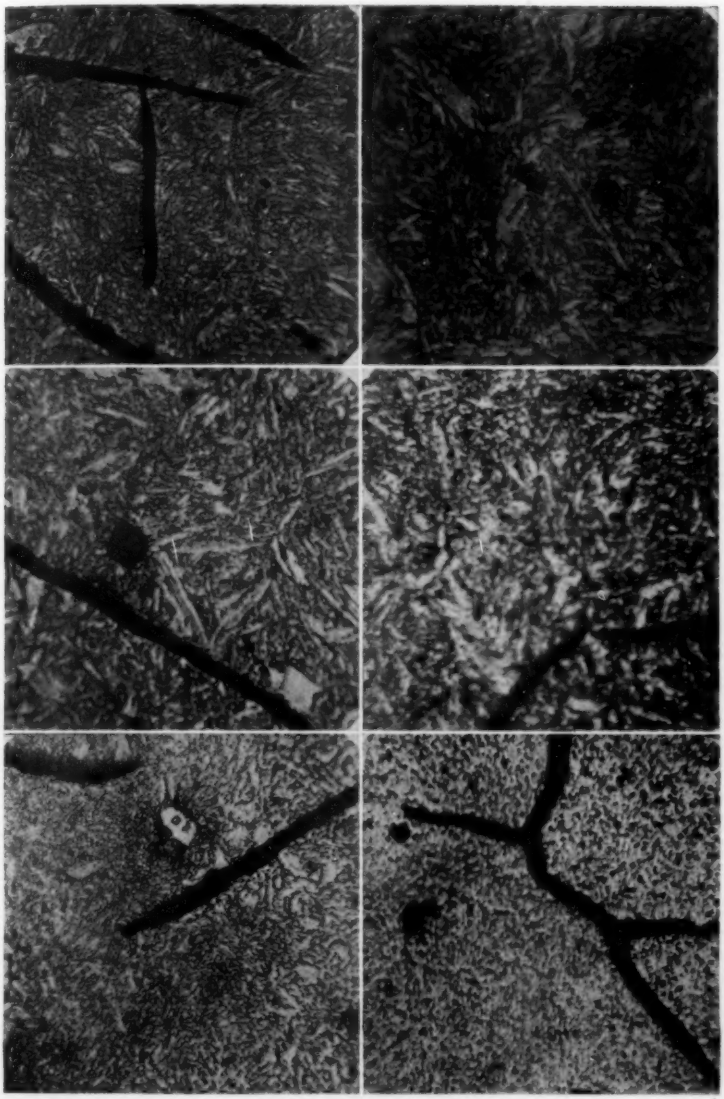


FIG. 10—BASE IRON (IRON NO. 1) QUENCHED FROM 1600°F. AND TEMPERED FOR ONE HOUR AT TEMPERATURES SHOWN. X1000. (UPPER LEFT)—TEMPERED AT 800°F. FIRTH HARDNESS 386. (UPPER RIGHT)—TEMPERED AT 900°F. FIRTH HARDNESS 310. (CENTER LEFT)—TEMPERED AT 1000°F. FIRTH HARDNESS 278. (CENTER RIGHT)—TEMPERED AT 1100°F. FIRTH HARDNESS 257. (LOWER LEFT)—TEMPERED AT 1200°F. FIRTH HARDNESS 221. (LOWER RIGHT)—TEMPERED AT 1300°F. FIRTH HARDNESS 172.

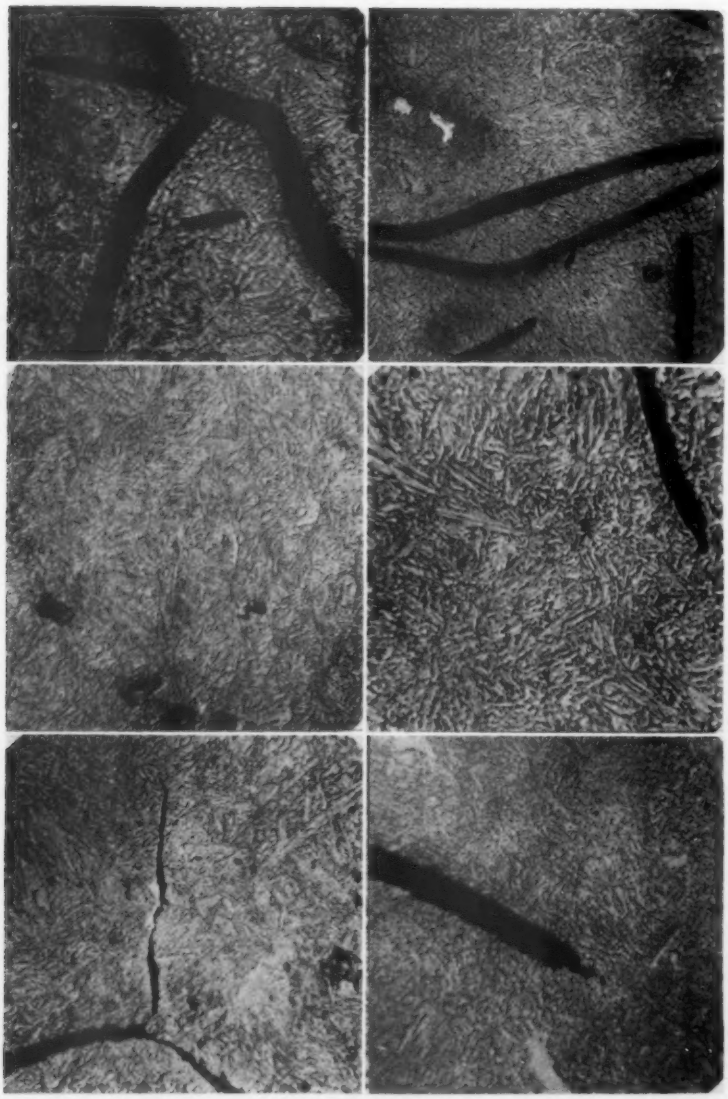


FIG. 11—ALLOYED IRON (IRON NO. 2) QUENCHED FROM 1600°F. AND TEMPERED FOR ONE HOUR AT TEMPERATURES SHOWN. x1000. (UPPER LEFT)—TEMPERED AT 200°F. FIRTH HARDNESS 573. (UPPER RIGHT)—TEMPERED AT 300°F. FIRTH HARDNESS 556. (CENTER LEFT)—TEMPERED AT 400°F. FIRTH HARDNESS 493. (CENTER RIGHT)—TEMPERED AT 500°F. FIRTH HARDNESS 490. (LOWER LEFT)—TEMPERED AT 600°F. FIRTH HARDNESS 481. (LOWER RIGHT)—TEMPERED AT 700°F. FIRTH HARDNESS 487.

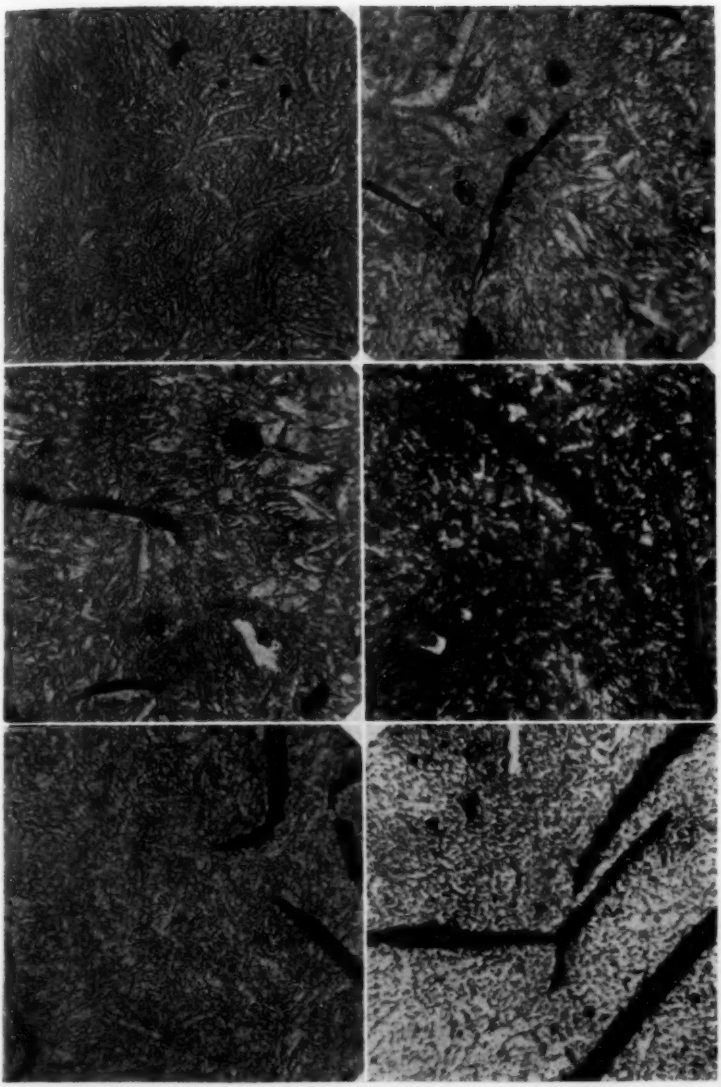


FIG. 12—ALLOYED IRON (IRON NO. 2) QUENCHED FROM 1600°F. AND TEMPERED FOR ONE HOUR AT TEMPERATURES SHOWN. x1000. (UPPER LEFT)—TEMPERED AT 800°F. FIRTH HARDNESS 425. (UPPER RIGHT)—TEMPERED AT 900°F. FIRTH HARDNESS 364. (CENTER RIGHT)—TEMPERED AT 1000°F. FIRTH HARDNESS 354. (CENTER LEFT)—TEMPERED AT 1100°F. FIRTH HARDNESS 349. (LOWER LEFT)—TEMPERED AT 1200°F. FIRTH HARDNESS 264. (LOWER RIGHT)—TEMPERED AT 1300°F. FIRTH HARDNESS 214.

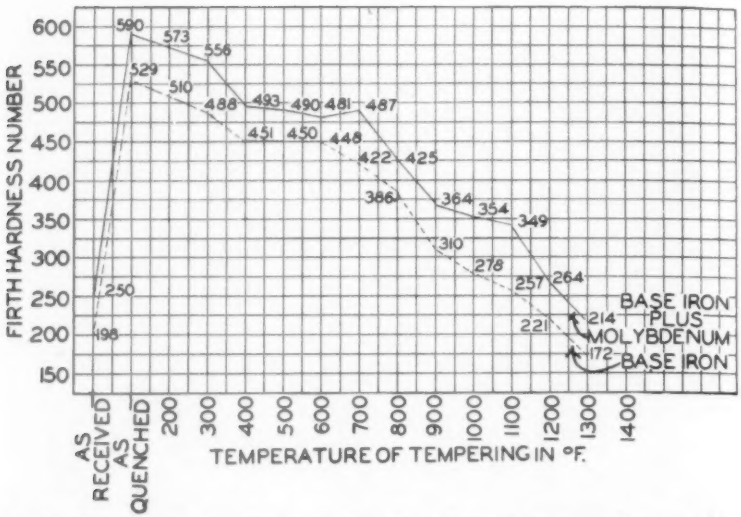


FIG. 13—HARDNESS OF BASE IRON AND BASE IRON PLUS MOLYBDENUM QUENCHED FROM 1600°F. AND TEMPERED FOR ONE HOUR AT TEMPERATURES SHOWN.

curve *B*, a finely pearlitic structure will be formed, and if cooling progresses along curve *C*, a coarse pearlitic structure will be produced.

36. In carrying out the isothermal transformation, the specimen is quenched from above the eutectoid temperature to some subcritical temperature, and held there for a length of time sufficient for the transformation to be complete. As an example, assume, on the hypothetical isothermal transformation curve, that the cooling rate be represented by curve *A*, down to temperature of 300°F. (Fig. 14). It is evident that at this temperature the structure will be completely austenitic. Fig. 14 shows that it takes about 4 sec. to cool the specimen to 300°F., and the iron remains austenitic until 30 sec. have elapsed. At this time, the austenite begins to transform to its transformation product, and the curve shows that the transformation is not complete until about 100,000 sec. have elapsed. At the end of the time period, the decomposition product of the austenite at this temperature, which is martensite, is formed. If a transformation temperature of 500°F. were chosen, the product of transformation would have been an acicular troostite. Similarly, the transformation product at 1000°F. is fine pearlite, and at 1300°F., a coarse pearlite.

37. In this hardness investigation, the authors have chosen

to keep the time of transformation a constant, and varied the transformation temperature, and then compared the hardness at the end of the given time period. This would be analogous to drawing a vertical line on Fig. 14 at 3600 sec. Under such conditions, it is evident that in many cases the transformation is completed, and in others it is not. In this investigation, the authors compared the products of transformation along the 3600 sec. line. To make this work complete, it would have been desirable to determine isothermal transformation curves for the irons studied, but time would not permit.

CONCLUSIONS

38. From the results obtained in this investigation, the authors draw the following conclusions:

1. On the isothermal transformation of the base iron and the alloyed molybdenum iron, the hardness curves are parallel over a temperature range of 200 to 1300°F. The difference in hardness between the two irons approaches a con-

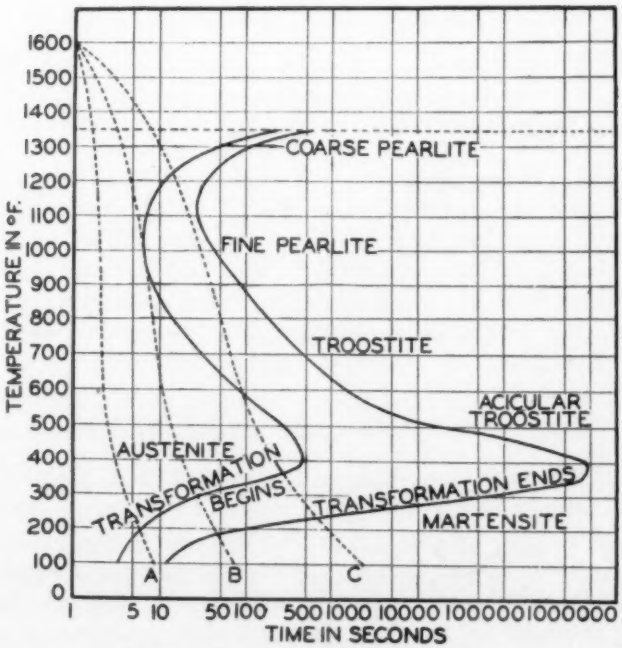


FIG. 14—THEORETICAL ISOTHERMAL TRANSFORMATION CURVE FOR CAST IRON.

stant value over the temperature range studied, after one hour of transformation.

2. On the quenching and tempering treatment, the alloyed molybdenum iron and the base iron showed hardness values that parallel each other over the range of 200 to 1300°F. The difference in hardness, after tempering a quenched specimen, approached a constant value in comparing the alloy and the base iron.

3. The addition of molybdenum to cast iron delays the rate of austenite transformation and increases the hardness of the samples after one hour of isothermal transformation.

4. The hardness of the quenched and tempered base iron, after a one hour temper, is higher than that produced on the same iron by isothermal transformation for the same length of time at the same temperature.

5. The hardness of the quenched and tempered alloyed molybdenum iron, after a one hour temper, is higher than that produced on the same iron on isothermal transformation for the same length of time at the same temperature.

6. The isothermal transformation of cast iron makes it possible to harden one inch sections of cast iron without having them crack.

7. In carrying out the isothermal transformation, the cast iron is quenched from 1600°F. to the desired transformation temperature in a salt bath, and then can be removed to an oven, at the same temperature, and the transformation allowed to go to completion. This would allow sufficient capacity for production hardening by the isothermal transformation method.

ACKNOWLEDGMENT

39. The authors are indebted to Dr. A. J. Herzig, chief metallurgist, Climax Molybdenum Co., Detroit, for supplying the irons used in this investigation; to Herbert F. Scobie, instructor, University of Minnesota, Mechanical Engineering Department, for analysis of the cast irons and to Dr. H. S. Jerabek and Dr. Arthur Forsyth for helpful suggestions during the course of the investigation.

DISCUSSION

Presiding: V. A. CROSBY, Climax Molybdenum Co., Detroit, Mich.

D. V. LUDWIG[1]: I feel I have a slight contribution to make to this paper. We ran into a threatened tie-up in our company on a matter of some 50-lb. ballistic cams. They were too porous and soft to be applied for the specific purpose intended. The cams were in various stages of machining having dimensional tolerances in some sections of such nature that losses in excess of 0.001-in. would not be tolerated. In much the same way, as outlined in the paper under discussion, we worked out a heat treatment whereby we were enabled to obtain constant and controlled hardness values, and as results later showed, a more desirable distribution of the graphitic carbon with a considerable refinement in porosity. We found that by heating the casting for an hour at 1550° F., quenching in oil and drawing as quickly as possible in air at times and temperatures ranging from 2 hrs. at 1000° F. for castings below 390 Brinell as quenched to 4 hrs. at 1100° F. for castings around 490-520 Brinell as quenched, we were enabled to secure hardnesses within the desired range of from 200 to 230 Brinell which was the intended base hardness of the material. Our Brinell figures were obtained on a basic Brinell machine and checked by Rockwell conversions.

The method of determining the hardness curve was very similar, except for the method of quenching, to the method outlined in the paper, even to the extent of the size and shape of the test specimens. I thought it might be of value to those present to know we were employing this procedure as a definite production control.

CHAIRMAN CROSBY: What hardness were those castings before you attempted to heat treat them?

MR. LUDWIG: That was the trouble we were confronted with. They were from 150 to 165 Brinell where we wanted 200 or better.

Table 3 showing the analysis secured from a bad casting in comparison with desired results may be of some interest. Except for the absence of nickel there are no real distinctions between the two analyses. The problem was one of mechanical distribution of the graphitic carbon which we solved by heat treatment.

T. E. EAGAN[2] (*Written Discussion*): From the analysis given it is assumed that the authors of this interesting and instructive paper were using cast iron with no steel additions. Also that the silicon additions to the ladle were for an innoculation treatment to control the graphite distribution.

Referring to Fig. 13 covering the hardness obtained on the quenched and drawn speciments. In the case of both types of irons there is a plateau where there is very little change in hardness, namely, 400 to 600°

[1]Metallurgist, Sperry Gyroscope Co., Inc., New York, N. Y.
[2]Chief Metallurgist, Cooper-Bessemer Corp., Grove City, Pa.

Table 3

DESIRED PROPERTIES COMPARED WITH BAD CASTING

Chemical Composition, Per Cent

	Graphitic Carbon	*Combined Carbon*	*Total Carbon*	*Mn.*	*P.*	*Si.*	*S.*	*Ni.*
Desired.....	3.05-3.30	0.60-0.75	2.40-2.65	0.60-0.80	0.20 (Max.)	1.80-2.10	0.01 (Max.)	0.50 (Min.)
Bad Casting	3.13	0.50	3.63	0.67	0.15	2.11	0.106	0.0

Physical Properties

	Brinell Hardness	*Tensile Strength, Lb. per sq. in.*	*Transverse Strength, Lb. per sq. in.*	*Deflection in 2-in., Per cent*
Desired	196-220	52,000	3,000	0.30-0.32
Bad Casting.....	160	30,000	2,200	0.30

F. for the base iron and 400 to 700° F. for the molybdenum iron. This condition was not discussed by the authors.

With their permission I would like to offer, as additional confirming evidence, another curve which follows the same trend as the two shown in this paper, Fig. 13. The analysis of the iron used in preparing these graphs is shown in Table 4. This iron contained 65 per cent steel and was graphitized by the addition of 0.2 per cent calcium silicide to the ladle.

Table 4

ANALYSIS USED FOR PREPARING GRAPHS

Total Carbon, per cent	2.95
Graphitic Carbon, per cent	2.36
Combined Carbon, per cent	0.59
Silicon, per cent	1.52
Phosphorus, per cent	0.11
Sulphur, Per cent	0.08
Manganese, per cent	0.78

Test bars were cut to the same dimensions as those used by the authors. They were quenched in oil temperatures of 1450, 1500, 1525, 1550, 1575, 1650°F. after preheating at 1000°F. and then held in a salt pot at

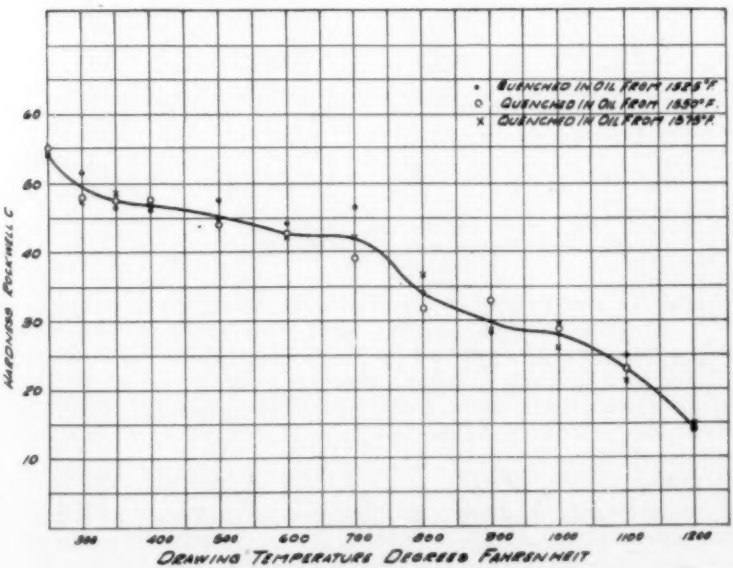
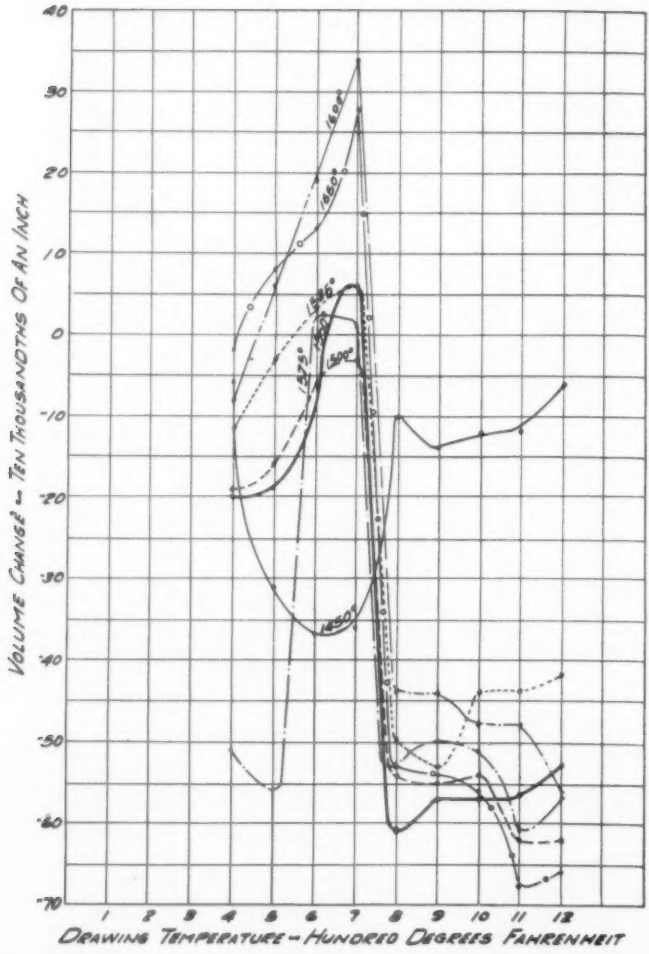


FIG. 15—EFFECT OF DRAWING TEMPERATURE ON HARDNESS.



the various temperatures mentioned for 35 min., then quenched in oil. Rockwell hardness readings were taken. The specimens quenched at 1525, 1550, and 1575°F. showed the best hardness readings, so these were subsequently drawn at various temperatures for 1 hr. The results are shown in Fig. 15. It can be readily seen that the general trend of this curve follows the trend of the two shown in Fig. 13. There is a definite plateau at 600 to 700°F.

In the study we fortunately made measurements of the test pieces and calculated the change in volume. All of the pieces were measured for this work and the change in volume is plotted in Fig. 16. Unfortunately we did not obtain the volume change below 400°F. It can be

seen that there is a definite shrinkage in volume up to 500°F. when the volume increases. The maximum increase being between 600 and 700°F. Above this temperature there is a very drastic shrinkage of the material. It has been found that this phenomenon takes place in the hardening and drawing of tool steels and it has been shown that this change is due to the transformation of retained austenite to martensite. Although the authors did not report any volume change figures, I am offering this as a suggested reason for the plateau that they show at around these same temperatures.

MR. DOWDELL: We are glad to have Mr. Eagan's contribution to our paper and agree with him fully that the transformation at 700°F. is austenite to martensite, or to be more technical, austenite to a tempered martensite.

Mr. Ludwig's contribution on the treatment of the 50-lb. ballistic cams also is very much appreciated and we wish to thank both Mr. Ludwig and Mr. Eagan for their contributions.

Some Properties of Synthetically Bonded Steel Molding Sands†

ROBERT E. MOREY* AND HOWARD F. TAYLOR*, ANACOSTIA, D. C.

Abstract

Synthetically bonded steel molding sands are studied to find the effect of changes in sand grain size, size frequency distribution and grain shape on the physical properties of the mixtures. Several clays were tested and their green compressive strengths compared. The clays vary considerably in strength and each clay has a green strength-water content curve which is characteristic of the clay. A widely used cereal binder of the corn flour type is studied in mixes with bentonite and in clay free mixes.

INTRODUCTION

1. Although metal castings have been made for centuries, the only method of evaluating the molding sand was by feeling the texture and strength with the hand. It is only recently that methods have been developed to measure accurately the properties of molding sands. Such work has been sponsored in this country since 1918 by the American Foundrymen's Association. In 1923 and 1924, the permeability test was developed and adopted as a tentative standard method. In 1925 some of the early compressive strength tests were made and some interesting work on them has been done by Dietert[1], Hansen[2] and Adams[3]. Prior to this time practically all molding sands were naturally bonded. That is, the sand and clay which they contained were combined in the proportions found in nature. Recently, however, synthetically bonded sands have come into widespread use. These consist of a silica base sand, often washed and graded, which is mixed with a suitable clay.

---

† Published by permission of the Navy Department.

* Sand Technologist and Associate Metallurgist, respectively, U. S. Naval Research Laboratory.

[1]Superior numbers refer to the bibliography at the end of this paper.

NOTE. Presented at a session on Foundry Sand Research at the 45th Annual A.F.A. Convention, New York City, May 13, 1941.

2. The reasons for the use of synthetically bonded sands are many but they all lead to the same end; they reduce the cost of the castings. The materials used are readily available while good natural sands are sometimes hard to obtain. Synthetically bonded sands are easy to control and, when properly used, they peel well and produce clean, smooth casting surfaces which reduces cleaning costs. Grain size and fines are easily controlled and there is little or no spent sand to haul away. The first cost is sometimes higher than that of natural sands but the durability of the synthetically bonded sands is so much greater that the cost per finished casting is less. Synthetic sands are well adapted to use in modern molding machines and continuous production systems. The physical properties can be adjusted to produce all classes of work with one or two base sands, whereas it is necessary to maintain large stocks of several natural sands to meet the varied needs of the foundry.

3. In previous reports at this laboratory, the effects of both water and bentonite contents on the physical properties of synthetically bonded steel molding sands were established. The principal facts brought out by this work are summarized in reference 9

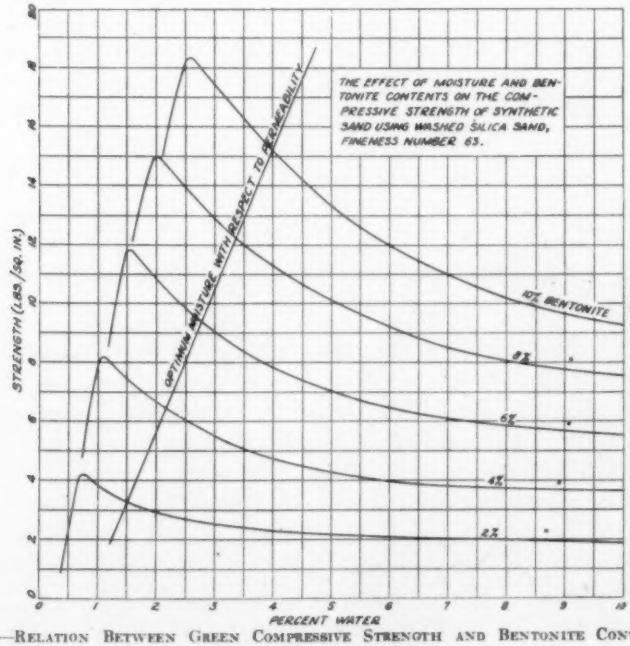


FIG. 1—RELATION BETWEEN GREEN COMPRESSIVE STRENGTH AND BENTONITE CONTENT OF STEEL FOUNDRY SANDS.

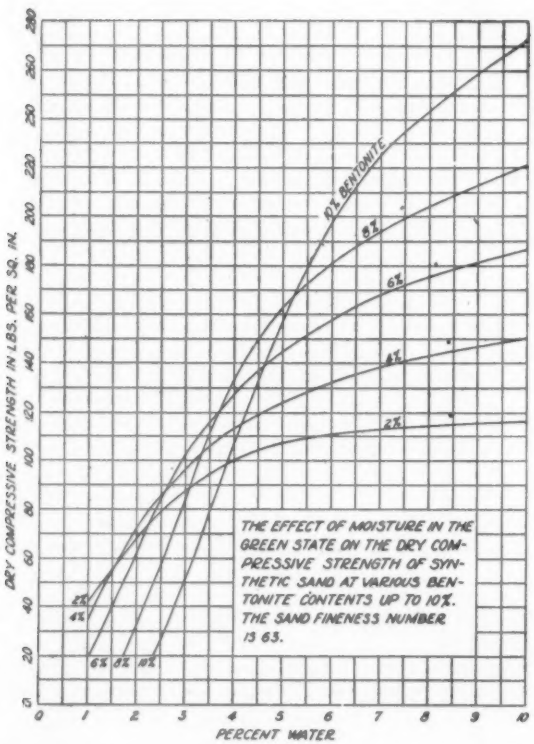


FIG. 2—DRY COMPRESSIVE STRENGTHS PRODUCED BY THE SAME MIXTURES SHOWN IN FIG. 1.

and are reproduced here as Figs. 1, 2 and 3. Figure 1 shows the relation between green compressive strength and bentonite and water contents. This graph proved to be of practical value for the control of green sand mixtures.

4. Since sand strength varies rapidly with both bentonite and water content, it is the usual practice to decide on values of green compressive strength and water content which are appropriate for the type of castings being produced. Plotting these values on the graph locates a point. The position of the point with respect to the bentonite curves indicates how much bentonite is required to produce the desired strength. In heap or used sands, where the strength of some of the bentonite is destroyed, it must be replaced if the properties of the sand are to be maintained constant. This is accomplished by making routine tests for green compressive strength and water content. Plotting these values on the graph will

indicate how much bentonite is necessary to bring the mixture back to its original strength.

5. Figure 2 shows the dry compressive strengths produced by the same mixes shown in Fig. 1. Figure 3 contains the same series of curves as Fig. 1 with an additional set superimposed. The additional curves were derived from the dry compressive strength curves. In this manner, it is possible to show the interrelation among the four variables, *i.e.*, water and bentonite contents, and green and dry compressive strengths. When any two of these are known, the other two may be found from the graph.

6. It is clear that these statements apply only to the sand and bentonite used for the tests. Other sands and other clays will give somewhat different results but the general character of the curves will be the same. It is a simple procedure to establish similar curves for any combination of sand and clay and since this work was presented, many foundrymen have stated their need for more information. They wish to know, for example, what changes in

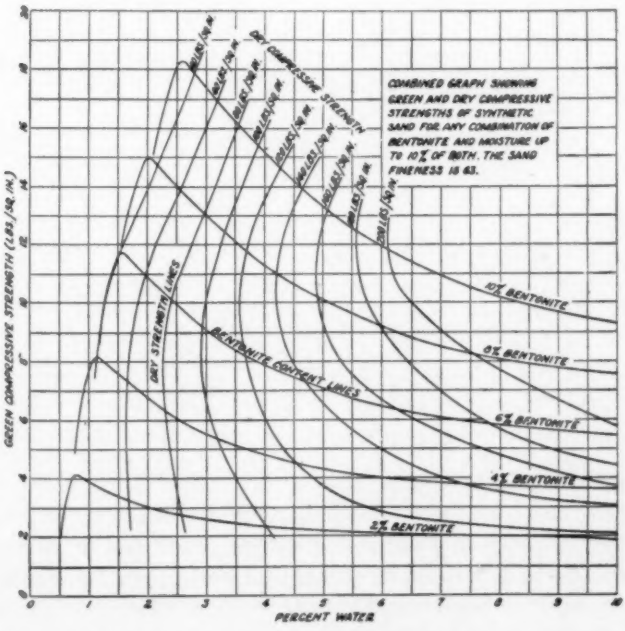

COMBINED GRAPH SHOWING GREEN AND DRY COMPRESSIVE STRENGTHS OF SYNTHETIC SAND FOR ANY COMBINATION OF BENTONITE AND MOISTURE UP TO 10% OF BOTH. THE SAND FINENESS IS 63.
DRY COMPRESSIVE STRENGTH
DRY STRENGTH LINES
BENTONITE CONTENT LINES
10% BENTONITE
8% BENTONITE
6% BENTONITE
4% BENTONITE
2% BENTONITE
GREEN COMPRESSIVE STRENGTH (LBS./SQ. IN.)
PERCENT WATER

FIG. 3—COMBINED GRAPH SHOWING GREEN AND DRY COMPRESSIVE STRENGTHS OF SYNTHETIC SAND FOR ANY COMBINATION OF BENTONITE AND MOISTURE UP TO 10 PER CENT OF BOTH.

the curves would be brought about by variations in sand grain size, size frequency distribution, grain shape and type of clay.

7. Information is needed also on cereal binders and their possible effect in controlling the tendency of bentonite bonded sands to dry out on the surface. Naturally bonded sands or washed sands bonded with fire clay do not exhibit this tendency. It might be possible to reduce surface drying by mixing other clays with bentonite. Grim[4] has reported that clays, such as kaolinite or illite containing small amounts of montmorillonite, have a strength out of all proportion to the amount of montmorillonite present. This suggests the possibility of utilizing the high strength of bentonite (composed chiefly of montmorillonite) and at the same time, by the addition of other clays, lessening surface drying.

METHODS USED IN THE INVESTIGATION

8. The sand testing methods used in this investigation are those recommended by the American Foundrymen's Association [10].

9. Test batches were made up so that the total weight of the dry materials was 2000 grams. Clay, cereal binder and sand contents are reported on the basis of this dry weight. The dry materials were placed in an 18-in. muller-type laboratory mixer and mixed dry for one minute. Water was then added and mixing continued for 5 min. The batch was dumped into a clean pail and transferred to half-gallon glass jars which were immediately tightly sealed. The sand was allowed to tempeer in the sealed jars for at least 24 hours. At the end of this time, the sand was removed, passed quickly through a ¼-in. sieve and returned to the jar. 100 grams of sand were then weighed out into a glass crystallizing dish for a moisture test. The sample was dried to constant weight in an oven at 105 to 110° C., cooled to room temperature in a desiccator, and reweighed. The loss of weight in grams was taken as the water content. It should be noted that the water content is based on total or wet weight and the percentages of dry materials on the dry weight.

10. To avoid variations due to the effects of temperature and humidity, a series of from 6 to 8 batches was made up at one time and allowed to temper. The jars were then opened and the samples screened and tested one at a time in as rapid succession as possible. This procedure was found to give excellent agreement in test values.

11. The standard green permeability, green compressive strength, dry permeability and dry compressive strength tests were made as well as tests for flowability, deformation and resilience. A careful check was kept on the amount of sand necessary to produce a test specimen exactly 2-in. long and this was recorded as weight of specimen. Since the volume is constant, this is also a measure of the density of the rammed mixture.

Testing Procedure

12. Briefly the testing procedure was as follows: A weighed amount of sand was placed in a steel tube with an inside diameter of 2-in. This was placed in the usual ramming device consisting of a plunger and a weight. The weight was raised and dropped three times. If the length of the test cylinder was 2-in. $+$ 1/32-in., it was used. If the specimen was not within this limit, it was discarded and another specimen was made, using an amount of sand corrected to produce a 2-in. long specimen. The permeability of the rammed specimen was determined by means of a permeability apparatus. The orifice method was used for all tests after it had been found to be in agreement with the stop watch method. A hand-operated, pendulum-type testing machine was used for making the green compression, deformation and resilience tests. Specimens for the dry permeability and compression tests were baked in an A.F.A. standard core oven, at a temperature between 105 and 110° C., for at least 2 hours. This oven is provided with forced ventilation, a rotating table and thermostatic control. The specimens were kept in a desiccator until they were tested. Dry permeability was measured by means of a special dry permeability tube with a mercury seal around the sides of the specimen which caused the air to travel the full length of the specimen as in the green permeability test. Dry strength was measured on the above-mentioned testing machine with the high strength attachment or, for very strong specimens, on a large Southwark-Emery testing machine.

13. Although sand sizes are expressed in A.F.A. fineness numbers throughout this report, it is somewhat more instructive to plot and compare them on the basis of their cumulative sieve analyses. Differences in size, distribution, clay and silt content as well as any unusual or abnormal characteristics of the sand are easily seen from the cumulative curve. The Tyler Method[5], with slight modifications, was used for this purpose. First, the sands

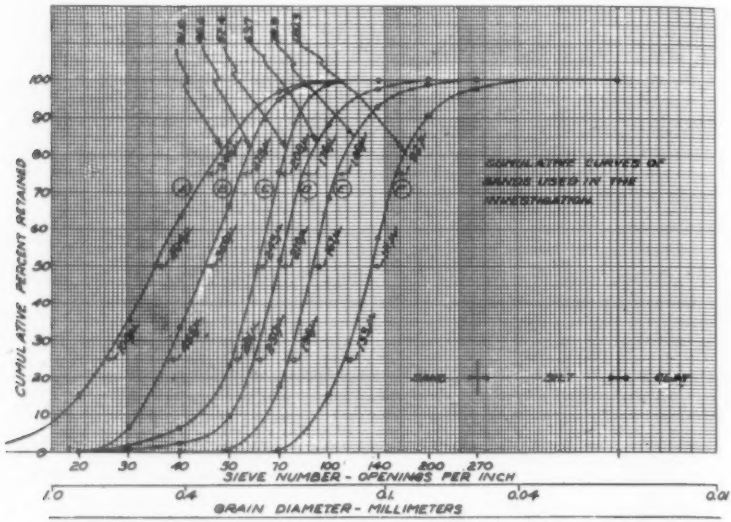


FIG. 4—CUMULATIVE CURVES OF ROUNDED GRAINED, WASHED SILICA SANDS *A*, *B*, *C*, *D*, *E*, *F*, USED IN THE INVESTIGATION. THESE SANDS HAVE APPROXIMATELY THE SAME GRAIN DISTRIBUTION BUT VARY IN SIZE.

were sieved through a set of 8-in. sieves. The cumulative percentages retained on the various sieves were then plotted as a function of the openings of the sieves in millimeters on a logarithmic scale. The resulting curves are shown in Figs. 4 to 6. The point where the cumulative curve crosses the 50 per cent retained line was noted and the size at this point is called "Center Grain Size." This may be expressed in millimeters or microns. In the same way, the diameters at 25 per cent and 75 per cent retained were found. The larger diameter divided by the smaller is called "Distribution Coefficient."

Discussion of Graph Papers

14. Two-cycle, semi-logarithmic, graph paper was used, although any semi-log paper will be satisfactory if the scales are appropriate. Three-cycle paper is especially suited to the molding gravels and naturally bonded molding sands while the two-cycle is usually ample for the synthetically bonded sands. Placing the paper so that the logarithmic scale decreases to the right, the end of the last cycle is given the value of 0.01 mm. The successive cycles to the left will then be 0.1 and 1 mm., respectively. The sizes in millimeters of the openings of the sieves are then marked on this scale and labeled with the mesh numbers of the sieves. The

linear scale of per cent retained is made on the left hand margin and the sieve analyses are then plotted as cumulative curves. To illustrate with an example, the 63.7 (A.F.A. fineness) sand was tested for fineness and plotted on Fig. 4. The curve was found to cross the 25, 50 and 75 per cent retained line at 0.250, 0.210, and 0.176 millimeters, respectively. The "center grain size" is therefore 0.210 mm. or 210 microns (microns = millimeters x 1000) and the "distribution coefficient" is

$$\frac{0.250}{0.176} = 1.42$$

Krumbein[6] has discussed the advantages of cumulative curves over size frequency curves. For the foundryman the most important advantages are:

1. Any sieves may be used regardless of the size, interval, or make, provided the size of the openings is known and is appropriate for the material being tested.

2. As few or as many sieves as desired may be used. Four or five may be ample for routine work, while 20 or 30 may be used for extreme accuracy.

3. All sieve analyses plotted in this way are comparable.

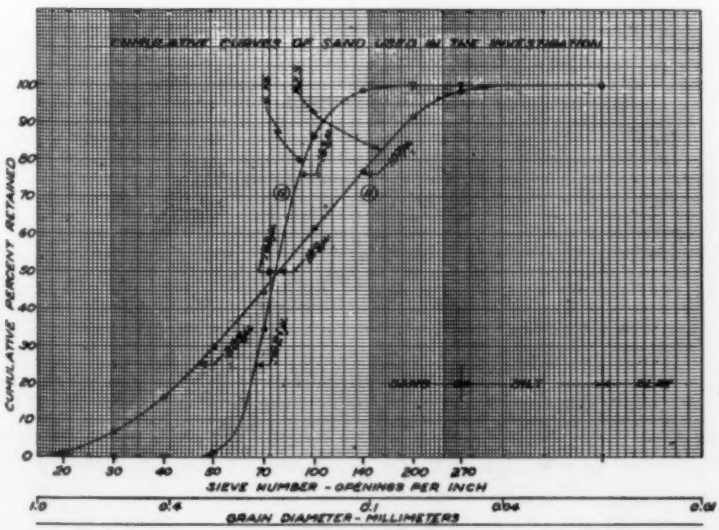


FIG. 5—CUMULATIVE CURVES OF TWO ROUNDED GRAINED, WASHED SILICA SANDS *G* AND *H* OF ABOUT THE SAME AVERAGE SIZE BUT DIFFERING WIDELY IN DISTRIBUTION.

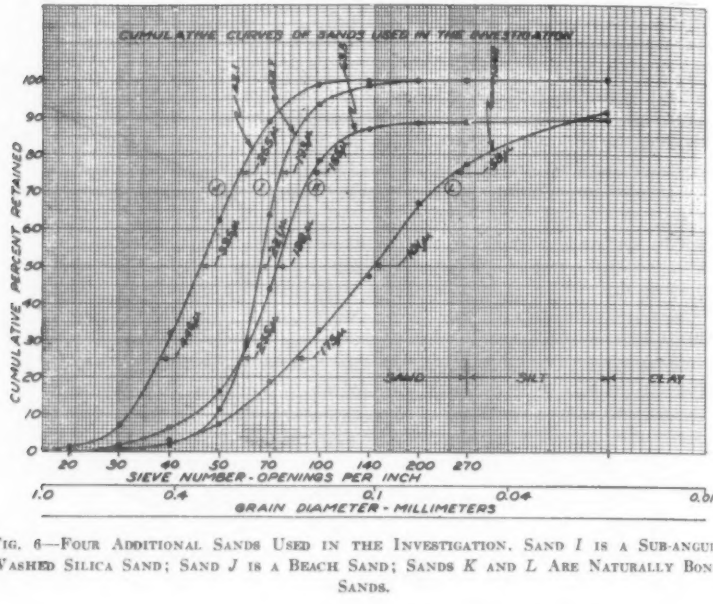


FIG. 6—FOUR ADDITIONAL SANDS USED IN THE INVESTIGATION. SAND *I* IS A SUB-ANGULAR, WASHED SILICA SAND; SAND *J* IS A BEACH SAND; SANDS *K* AND *L* ARE NATURALLY BONDED SANDS.

American sands separated on either Tyler or U. S. Standard sieves may be compared with British or Continental sands.

MATERIALS USED IN THE INVESTIGATION

15. Sieve analyses and cumulative curves of the sands used are shown in Table 1, and Figs. 4 to 6, inclusive. Sands *A* to *F*, inclusive, are washed silica sands with rounded grains, all of which have approximately the same distribution but vary in size. These sands were used in the study of the effect of grain size on the physical properties of the mixture.

16. Sands *G* and *H* are also washed silica with rounded grains. These are of about the same average size but differ widely in distribution, sand *G* having grains of a very uniform size and sand *H* having grains of many sizes.

17. Sand *I* is a washed silica sand but has sub-angular instead of rounded grains. Sand *J* is a beach sand from Cape Henry and has angular grains.

18. Sand *K* is a Downer steel molding sand and *L* is an Albany sand. A comparison of the test results of these sands with

SIEVE ANALYSES OF SANDS USED IN THE INVESTIGATION

Mesh	Per Cent Retained	Sand A Indi-vid.	Sand A Cumul.	Sand B Indi-vid.	Sand B Cumul.	Sand C Indi-vid.	Sand C Cumul.	Sand D Indi-vid.	Sand D Cumul.	Sand E Indi-vid.	Sand E Cumul.	Sand F Indi-vid.	Sand F Cumul.
Retained on 6 mesh		...	...	...	...	...	...	...	...	...	...	...	...
" 12 "		0.20	0.20	...	...	...	...	...	...	...	...	...	...
" 20 "		14.80	15.00	0.16	0.16	0.40	0.40	0.14	0.14	...	...	0.02	0.02
" 30 "		20.44	35.44	6.44	6.60	1.34	1.74	0.44	0.58	...	...	0.02	0.04
" 40 "		28.14	63.58	27.02	33.62	4.64	6.38	1.72	2.30	0.02	0.02	0.02	0.06
" 50 "		20.24	83.82	32.62	66.24	16.76	23.14	7.54	9.84	0.80	0.82	0.02	0.08
" 70 "		13.16	96.98	28.40	94.64	52.34	75.48	41.50	51.34	16.86	17.68	0.28	0.36
" 100 "		2.78	99.76	4.86	99.50	23.18	98.66	36.76	88.10	50.70	68.38	14.80	15.16
" 140 "		0.34	100.10	0.50	100.00	1.48	100.14	9.36	97.46	24.60	92.98	42.12	57.28
" 200 "		0.02	100.12	0.06	100.06	0.10	100.24	2.56	100.02	6.00	98.98	33.62	90.90
" 270 "		...	100.12	...	100.06	0.02	100.26	0.30	100.32	0.90	99.88	6.56	97.46
Silt		...	100.12	...	100.06	0.04	100.30	0.12	100.44	0.34	100.22	2.44	99.90
Clay		...	100.12	...	100.06	...	100.30	...	100.44	...	100.22	...	99.90
A.F.A. Fineness		31.0		40.6		52.4		63.7		79.8		120.3	
Center Grain Size, Microns		494		350		243		210		167		111	
Distribution Coef.		2.03		1.67		1.40		1.42		1.40		1.45	

Mesh	Per Cent Retained	Sand G Indi-vid.	Sand G Cumul.	Sand H Indi-vid.	Sand H Cumul.	Sand I Indi-vid.	Sand I Cumul.	Sand J Indi-vid.	Sand J Cumul.	Sand K Indi-vid.	Sand K Cumul.	Sand L Indi-vid.	Sand L Cumul.
Retained on 6 mesh		...	...	...	...	...	...	...	...	...	...	...	...
" 12 "		...	...	...	...	...	...	0.02	0.02	...	...	...	...
" 20 "		...	...	0.80	0.80	0.16	0.16	0.98	1.00	0.40	0.40	0.58	0.58
" 30 "		0.02	0.02	5.40	6.20	0.38	0.54	6.02	7.02	1.20	1.60	0.58	1.16
" 40 "		0.05	0.07	10.19	16.39	1.30	1.84	24.90	31.92	4.54	6.14	1.86	3.02
" 50 "		1.00	1.07	13.58	29.97	9.64	11.48	30.20	62.12	9.86	16.00	4.20	7.22
" 70 "		33.74	34.81	14.51	44.48	52.00	63.48	26.82	88.94	27.82	43.82	11.58	18.80
" 100 "		51.60	86.41	17.00	61.48	29.94	93.42	10.02	98.96	34.58	78.40	13.64	32.44
" 140 "		12.42	98.83	15.32	76.80	5.20	98.62	0.82	99.78	8.44	86.84	14.66	47.10
" 200 "		1.38	100.21	14.75	91.55	1.30	99.92	...	99.78	1.58	88.42	19.82	66.92
" 270 "		0.03	100.24	7.27	98.82	0.18	100.10	...	99.78	0.40	88.82	10.40	77.32
Silt		0.01	100.25	1.19	100.01	0.06	100.16	...	99.78	0.46	89.38	14.04	91.36
Clay		...	100.25	...	100.01	...	100.16	...	99.78	10.72	100.00	8.64	100.00
A.F.A. Fineness		67.7		82 9		58.7		42.1		63.3		134.8	
Center Grain Size, Microns		192		189		221		335		198		101	
Distribution Coef.		1.36		3.11		1.32		1.75		1.65		3.02	

those above serves to bring out some of the differences between naturally bonded and synthetically bonded sands.

19. The clays used were a Wyoming bentonite, composed chiefly of the mineral montmorillonite; an Illinois clay, composed chiefly of the mineral illite; and a Pennsylvania fire clay. The Downer and Albany sands contained their natural clays. Their mineral composition was not determined.

EFFECT OF GRAIN SIZE ON PROPERTIES OF SYNTHETICALLY BONDED SANDS

20. Each of the six sands, *A* to *F*, inclusive, in Table 1, was used as the base sand for a series of mixtures. Each series contained 5 per cent bentonite with moisture varying between 1 and 10 per cent. The physical properties of each mix were measured as described above and the results are shown on Figs. 7 to 12, inclusive. As can be seen from the curves each property of the sand varies with water content. Because of the similarity of the characteristic curves showing variations in physical property as a function of water content, it is only necessary to discuss one set of

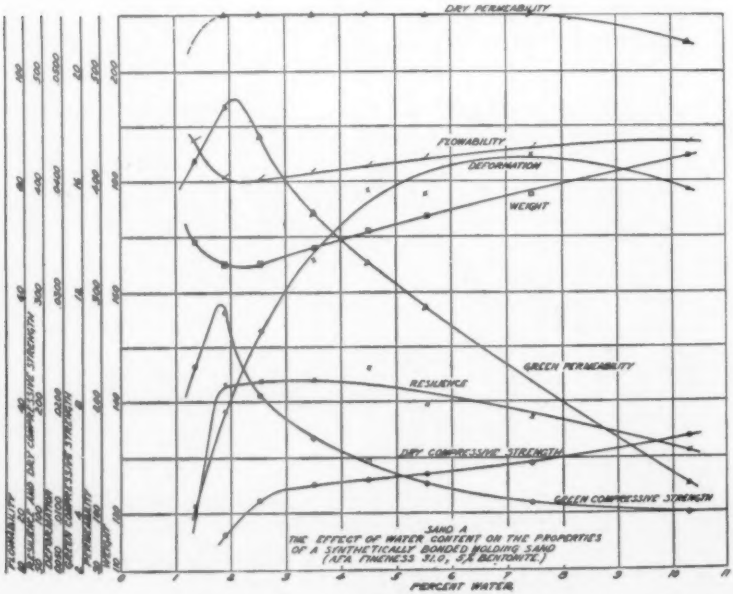


FIG. 7—EFFECT OF WATER CONTENT ON THE PROPERTIES OF WASHED SILICA SAND *A* BONDED WITH 5 PER CENT BENTONITE.

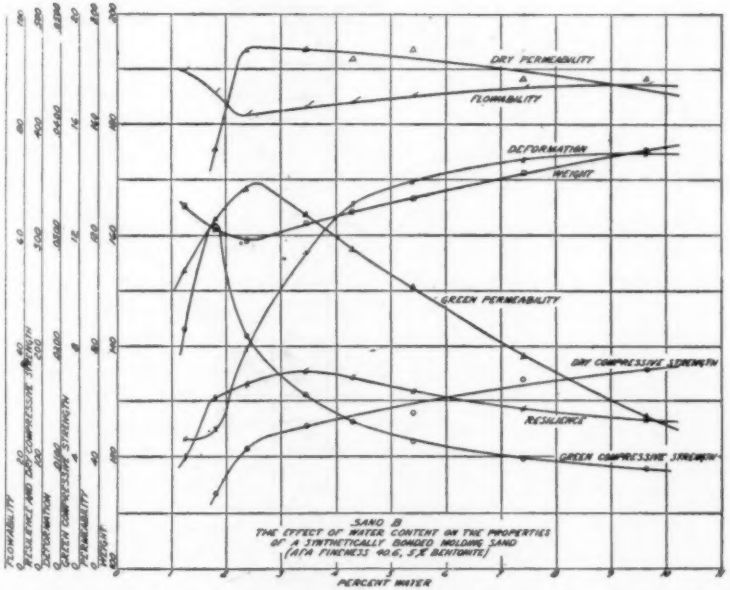


FIG. 8—EFFECT OF WATER CONTENT ON THE PROPERTIES OF WASHED SILICA SAND *B* BONDED WITH 5 PER CENT BENTONITE.

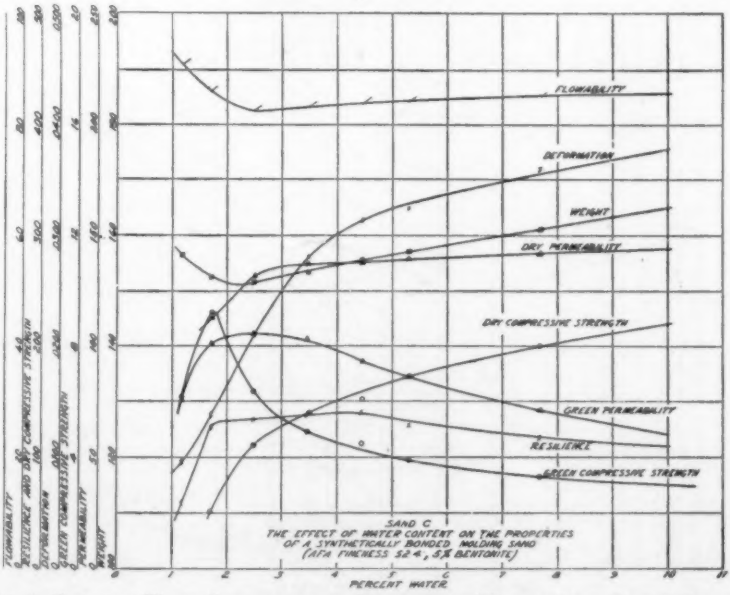


FIG. 9—EFFECT OF WATER CONTENT ON THE PROPERTIES OF WASHED SILICA SAND *C* BONDED WITH 5 PER CENT BENTONITE.

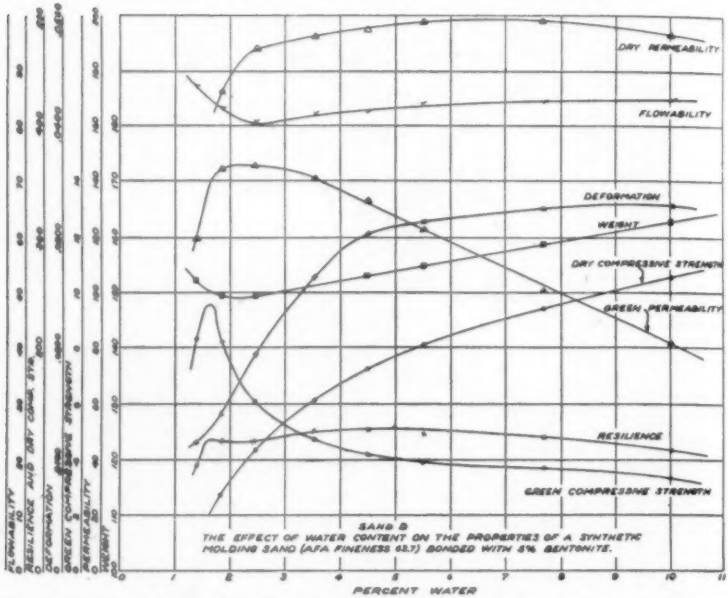


FIG. 10—EFFECT OF WATER CONTENT ON THE PROPERTIES OF WASHED SILICA SAND *D* BONDED WITH 5 PER CENT BENTONITE.

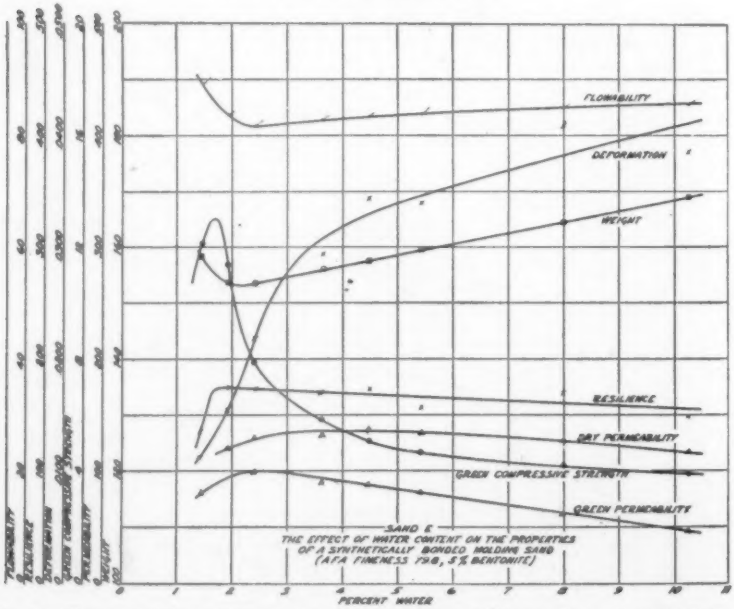


FIG. 11—EFFECT OF WATER CONTENT ON THE PROPERTIES OF WASHED SILICA SAND *E* BONDED WITH 5 PER CENT BENTONITE.

curves in detail. Figure 10 was selected for this purpose because these mixes contain sand of a size (A.F.A. fineness 63.7) which is very commonly used.

21. The batch containing 1.17 per cent water was so dry that it was very difficult to work and was definitely too dry for molding purposes. In this moisture range, the sand has a tendency to pack very tightly, as shown by the high weight and low permeability values. The green strength was 6.16 lb. per sq. in. In the next batch, the water was increased to 1.70 per cent. The increase in water "opened up" the sand and caused a large increase in permeability with a corresponding decrease in weight. The green strength was also increased from 6.16 to 9.25 lb. per sq. in. The third batch contained 2.48 per cent water which produced a small increase in permeability and a large decrease in strength, the weight being slightly reduced. Further increase in water content increased weight and decreased green compressive strength and green permeability. The position of the maximum and minimum points of the various curves change with different sands but their general shape is about the same for all the sands which were studied.

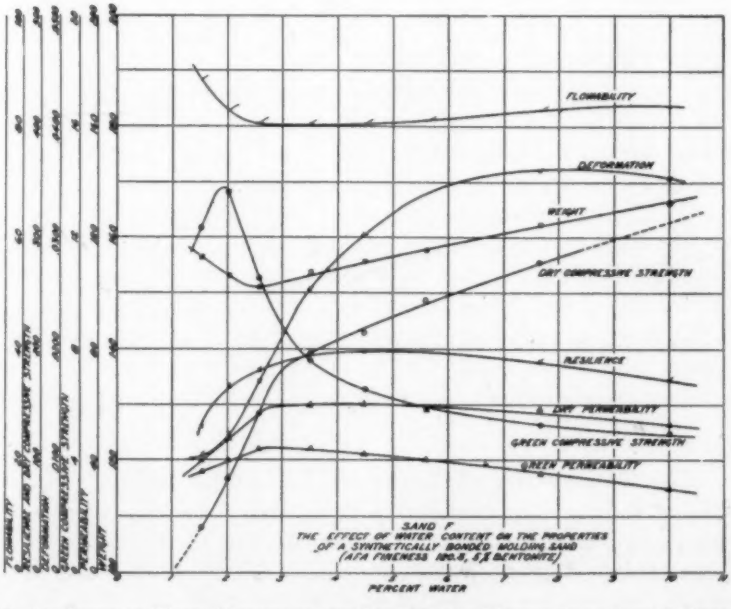


FIG. 12—EFFECT OF WATER CONTENT ON THE PROPERTIES OF WASHED SILICA SAND *F* BONDED WITH 5 PER CENT BENTONITE.

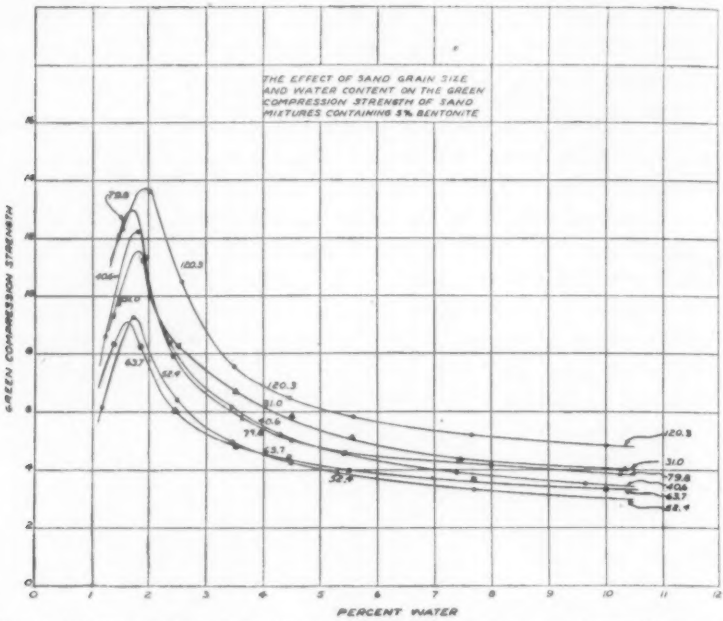


FIG. 13—EFFECT OF SAND GRAIN SIZE AND WATER CONTENT ON THE GREEN COMPRESSION STRENGTH OF WASHED SILICA SANDS BONDED WITH 5 PER CENT BENTONITE.

22. Flowability, deformation, and resilience curves also are shown but so far they have been of very little assistance in controlling sand used for molding at this laboratory. A minimum flowability of 82.5 per cent is indicated at 2.5 per cent water. This is well above the amount needed for satisfactory molding[7]. Deformation and resilience values are erratic, although in general they increase with increase in water content. Dry permeability for any given mix is always higher than green permeability. At low moisture contents, the difference is small and increases in proportion to the increase in moisture. Dry compressive strength increases rapidly with water content up to about 2 or 3 per cent and then increases more gradually as the water is increased to 10 per cent.

Compressive Strength

23. To show the effect of sand grain size on the properties of molding mixtures, certain curves from Figs. 7 to 12 have been selected and drawn as composite curves in Figs. 13 to 17. Figure 13 shows the effect of sand size on green compressive strength. This set of curves shows clearly the remarkable similarity in the shape of the strength curves for different sands. The maximum strength

occurs in each case between the limits of 1.6 and 1.9 per cent water. Since 5 per cent bentonite was used, this gives clay-water ratios from 2.6 to 3.1. In general, the highest strengths were obtained with the very coarse and very fine sands, the 31.0 and 120.3 sands (A.F.A. grain fineness numbers) being the strongest, and the intermediate sizes, 52.4 and 63.7, being the weakest of those tested. This relation is probably influenced by two factors, the surface area and the number of grain contacts. The coarse sands have a relatively small surface area as compared to the fine sands so that for a given percentage of clay, the grains are covered with a thick layer which tends to produce strength. The fine sands with a large surface area have a thin layer of clay but there are a much larger number of grain contacts which contribute to the strength.

Green Permeability

24. The green permeability curves are collected in Fig. 14. All the curves are similar in shape, increasing rapidly to a maximum and then decreasing approximately as a linear function of the water content. The maxima occur between 2 and 3 per cent water, regardless of sand size. The magnitude of the maximum

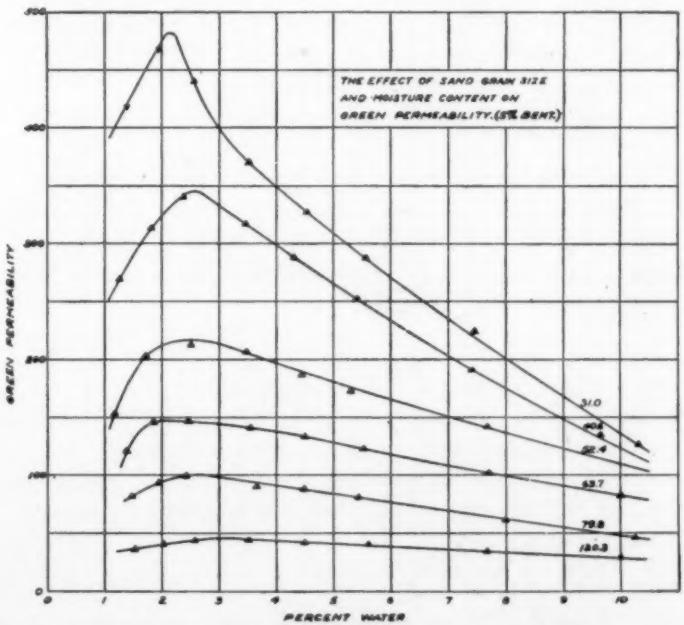


FIG. 14—EFFECT OF SAND GRAIN SIZE AND WATER CONTENT ON GREEN PERMEABILITY OF WASHED SILICA SANDS BONDED WITH 5 PER CENT BENTONITE.

naturally varies considerably, being 45 for the fine sand and over 470 for the coarse sand.

Dry Permeability

25. The dry permeability curves are shown in Fig. 15. Here again, increases in grain size produce corresponding increases in dry permeability. In most cases, the permeability increases to a

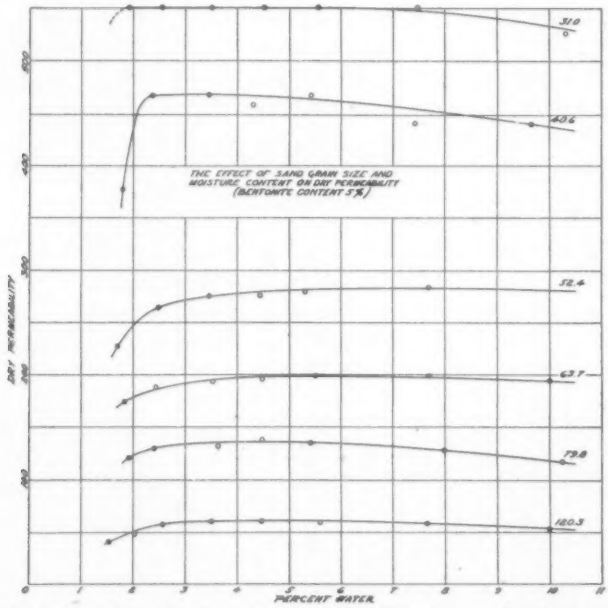


FIG. 15—EFFECT OF SAND GRAIN SIZE AND WATER CONTENT ON DRY PERMEABILITY OF WASHED SILICA SANDS BONDED WITH 5 PER CENT BENTONITE.

maximum with increase in water content and is nearly constant with further additions of water.

Dry Strength

26. As stated above, dry strength is very low at low moisture contents. It increases rapidly at first and then more gradually as the water content increases. This is shown by the curves assembled in Fig. 16. In general, the fine sands produce higher strengths. The 120.3 sand, for instance, is nearly twice as strong as the 31.0 sand.

Density

27. Figure 17 shows the effect of sand grain size and moisture content on the weight of the specimens which is a measure

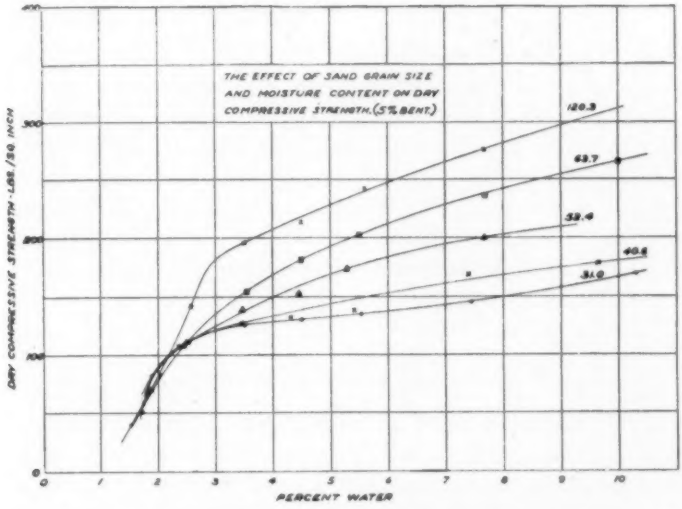


FIG. 16—EFFECT OF SAND GRAIN SIZE AND WATER CONTENT ON DRY STRENGTH OF WASHED SILICA SANDS BONDED WITH 5 PER CENT BENTONITE.

of the density of the mixture. The coarse sands (31.0 and 40.6) are the most dense while the remaining four are so nearly alike that it is difficult to draw any conclusion regarding them. The shape of the curve is the same for all the sands, with the minima occurring in every case between 2.0 and 2.5 per cent water.

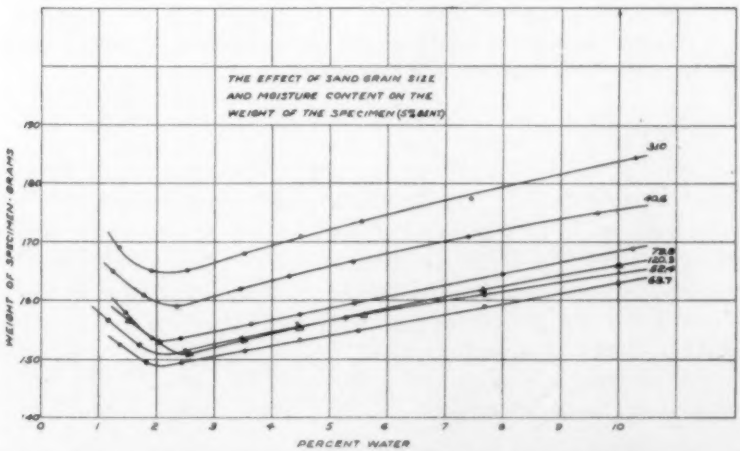

THE EFFECT OF SAND GRAIN SIZE AND MOISTURE CONTENT ON THE WEIGHT OF THE SPECIMEN (5% BENT.)
WEIGHT OF SPECIMEN - GRAMS
PERCENT WATER
31.0
40.6
63.7

FIG. 17—EFFECT OF SAND GRAIN SIZE AND WATER CONTENT ON SPECIMEN WEIGHT (DENSITY) OF WASHED SILICA SANDS BONDED WITH 5 PER CENT BENTONITE.

EFFECT OF GRAIN SHAPE

28. In an effort to determine the effect of grain shape on the physical properties of molding sands, two sands were tested—one having angular and the other sub-angular grains. Comparison of these two series of tests with each other and with the 63.7 sand (Fig. 10) failed to show any differences which could be attributed directly to grain shape. It is planned to study this variable further with suitable sands separated by sieves to obtain samples of the same grain size and distribution but having different shapes varying from round to angular.

EFFECT OF DISTRIBUTION

29. To find the effect of distribution of the base sand on the properties of a synthetic sand mixture, two sands were prepared from samples separated in an A.F.A. sieve series. They were mixed to give one sand very uniform in grain size and one sand widely distributed (Table 1 and Fig. 5). The sieve analyses and cumulative curves of these sands indicate that they are comparable to types normally used in synthetically bonded molding mixes. The slopes of the curves show the difference in the distribution of the two sands. Although the fineness numbers are not the same, it may be seen that both sands have approximately 50 per cent of the grain larger and 50 per cent smaller than 190 microns and that they are distributed symmetrically about this point. Six batches of each type were mixed in the usual way and tempered to water contents from 1.2 to 10 per cent. Figures 18 and 19 show the results of the tests on the two series.

Compressive Strength

30. In Figure 20, the green compressive strengths of the two sands are compared. The distributed sand is stronger over the full range of water contents but distribution has no effect on the shape of the curve or the position of the maximum strength with respect to water content. The two sands were about equally workable but the distributed sand appeared to have more resistance to rubbing off than the uniform sand.

Green Permeability

31. Figure 21 compares the green permeabilities. Here again the curves are almost identical in shape and position of the maxima

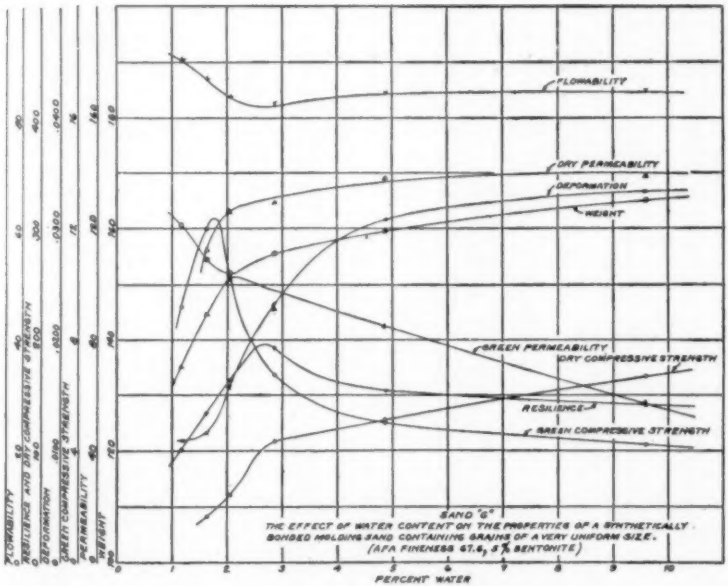


FIG. 18—EFFECT OF WATER CONTENT ON THE PROPERTIES OF A WASHED SILICA SAND CONTAINING 5 PER CENT BENTONITE. THE BASE SAND CONTAINED GRAINS OF EXTREMELY UNIFORM SIZE, SAND *G* OF TABLE 1.

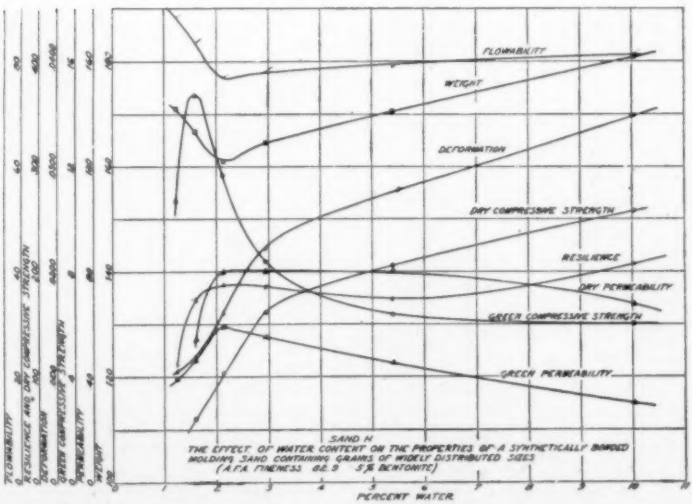


FIG. 19—EFFECT OF WATER CONTENT ON SAND *H*, TABLE 1, BONDED WITH 5 PER CENT BENTONITE. THIS SAND CONTAINS GRAINS OF WIDELY DISTRIBUTED SIZES.

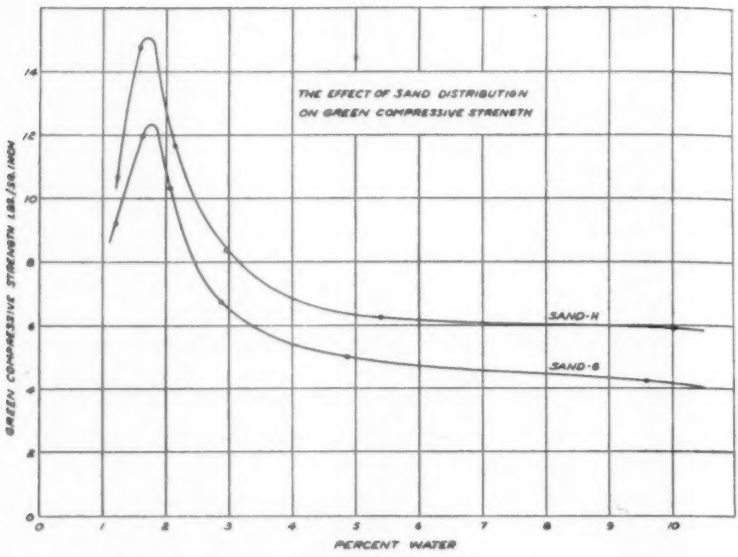


FIG. 20—EFFECT OF GRAIN DISTRIBUTION ON GREEN COMPRESSIVE STRENGTH.

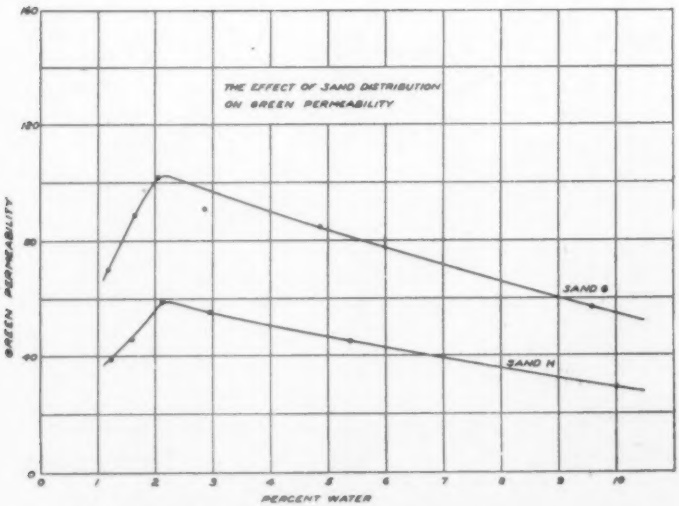


FIG. 21—EFFECT OF GRAIN DISTRIBUTION ON GREEN PERMEABILITY.

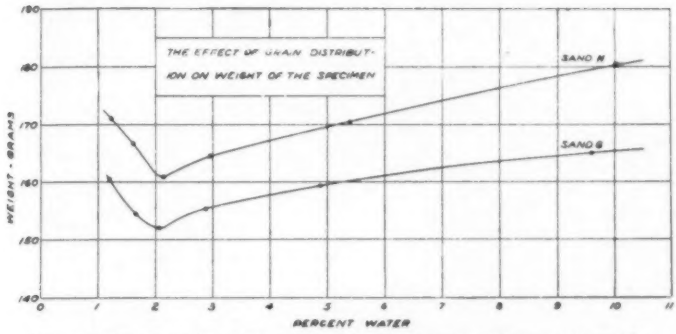


FIG. 22—EFFECT OF GRAIN DISTRIBUTION ON SPECIMEN WEIGHT (DENSITY).

with respect to water but the permeability of the distributed sand is only about 60 per cent of that of the uniform sand.

Density

32. Figure 22 shows a large difference in density between the two sands, the distributed sand being from 6 to 10 per cent more dense than the uniform sand. This is explained by the "fitting in" of small grains between the large ones[8] which reduces the voids and therefore decreases the permeability. These small grains also increase green strength by increasing the number of grain contacts.

Flowability

33. From Fig. 23, it is seen that uniformity of grain size increases flowability by about 5 per cent. Both sands, however, have ample flowability for all practical purposes.

Dry Strength

34. The dry strengths of these sands are shown in Fig. 24. Here the distributed sand shows a decided advantage. It is over 50 per cent stronger than the uniform sand at the same water con-

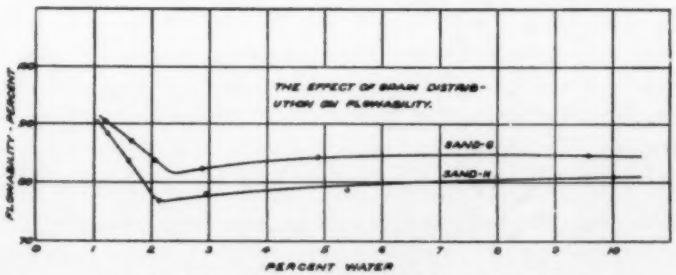


FIG. 23—EFFECT OF GRAIN DISTRIBUTION ON FLOWABILITY.

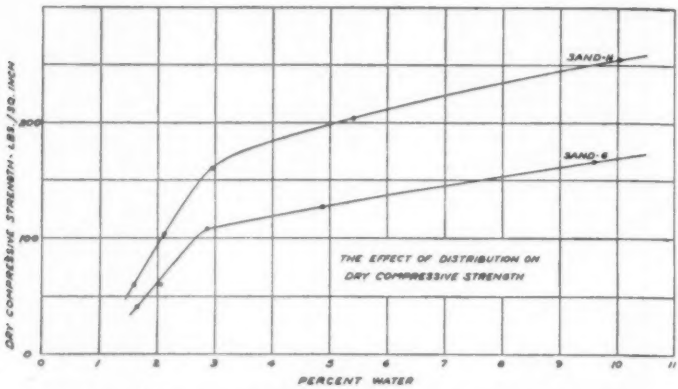


FIG. 24—EFFECT OF GRAIN DISTRIBUTION ON DRY COMPRESSIVE STRENGTH.

tent, or, to produce 180 lb. per sq. in. dry compressive strength requires 11 per cent water in the uniform sand but only 3.8 per cent in the distributed sand. Baked specimens of the two sands were equally good in surface finish but the distributed sand had more resistance to rubbing off and therefore would be less apt to wash in a mold.

35. In the past a uniform sand has been thought to be the best base for a synthetic molding sand. The curves shown here, however, indicate that a sand which is distributed over several sieves has many points of superiority over the uniform sand.

EFFECT OF TYPE OF CLAY

36. In all the sand mixes tested up to this point, bentonite was used as the bonding clay. Many other clays, however, are successfully used in the production of steel castings, fireclay being one of the most common. They must, however, be used in larger percentages than bentonite. Figures 25 and 26 show tests of two series made with the 63.7 sand and 10 and 20 per cent fireclay respectively. It will be seen that the green compressive strength curves have a shape which is characteristic of fireclay and that it varies somewhat from that produced by similar mixes containing bentonite. The maxima in the permeability curves occur at much higher water contents than is the case with bentonite bonded sands.

37. One of the most interesting differences between the fireclay and the bentonite bonded sand is in their behavior with respect to the ability of moisture to migrate more or less freely from place

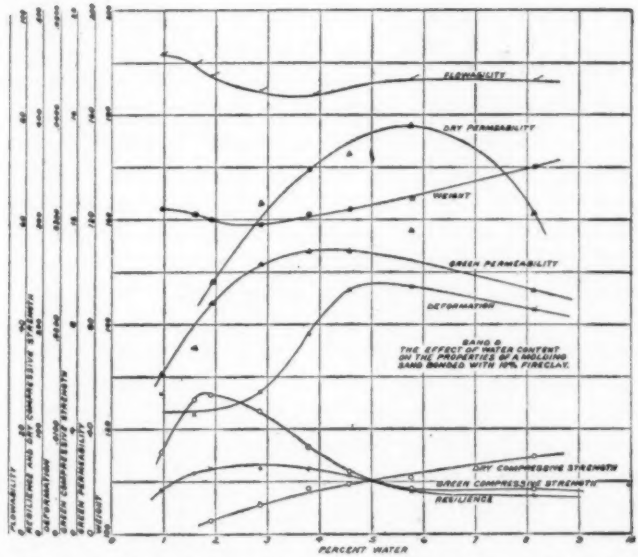


FIG. 25—EFFECT OF WATER CONTENT ON THE PROPERTIES OF A WASHED SILICA SAND (SAND *D*, TABLE 1) BONDED WITH 10 PER CENT FIRE CLAY.

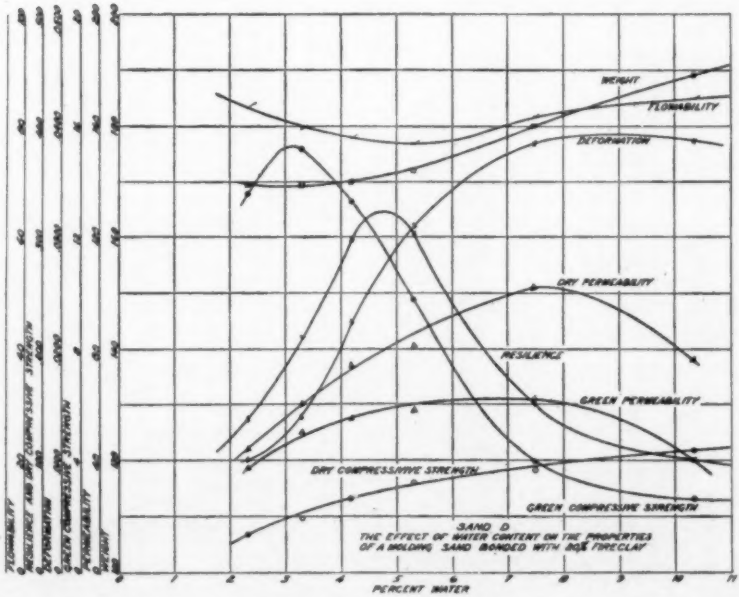


FIG. 26—EFFECT OF WATER CONTENT ON THE PROPERTIES OF WASHED SILICA SAND *D*, TABLE 1, BONDED WITH 20 PER CENT FIRE CLAY.

to place within the mold. In the bentonite sands, this migration is very limited; that is, if a completely saturated spot is next to a dry spot the moisture distribution remains practically unchanged. This characteristic was easily shown by dropping an equal amount of water on the tops of two test specimens, one made with bentonite and the other with fireclay, and, after several minutes, cutting them in half vertically. In the bentonite-bonded sand, the moisture remains near the top while in the fireclay-bonded sand it will be found distributed over a large part of the volume. This characteristic of bentonite leads to rapid surface drying, since the surface moisture is evaporated and is not replaced by moisture from the backing sand. Wetting agents have been tried in an effort to modify this characteristic of the bentonite but they have not been found to be very effective. It is possible that clays, such as bentonite and fireclay, can be blended to produce a mixture having the desirable properties of both clays. Too little work has been done to date, however, to admit of more than a general statement of this possibility.

38. To show how the properties of naturally bonded sands compare with synthetically bonded sands, a steel molding sand having a fineness number of 52.1 and 10.7 per cent clay was tested (Fig. 27).

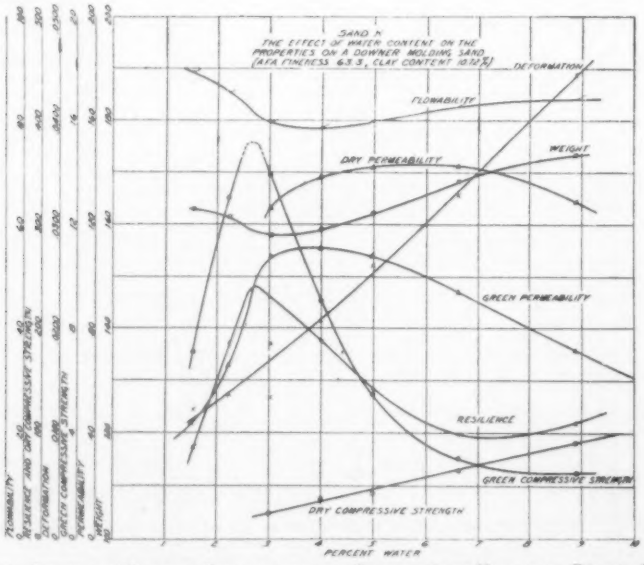


FIG. 27—EFFECT OF MOISTURE CONTENT ON THE PROPERTIES OF NATURALLY BONDED SAND *K*. TABLE 1, CONTAINING 10.72 PER CENT CLAY.

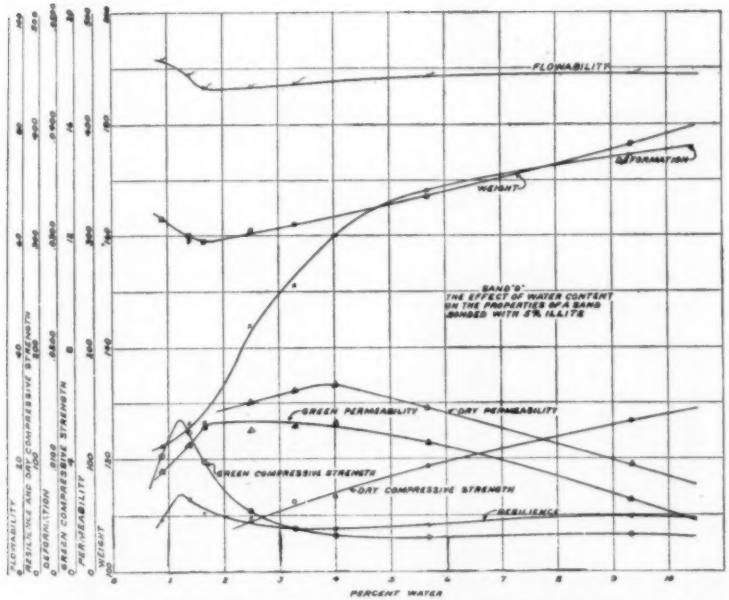


FIG. 28—EFFECT OF MOISTURE CONTENT ON THE PROPERTIES OF WASHED SILICA SAND *D* BONDED WITH 5 PER CENT OF AN ILLITE CLAY.

39. Figure 28 shows the physical properties of an Illinois clay kindly supplied by Dr. R. E. Grim, Illinois State Geological Survey. Its principal constituent is the clay mineral illite. A similar series, made with the same sand bonded with bentonite, whose principal constituent is the clay mineral montmorillonite, was shown in Fig. 10. A comparison of these series shows that the clays are very similar in effect. The illite gave approximately 40 per cent of the green strength produced by montmorillonite.

40. This is shown more clearly in Fig. 29, which gives strength curves produced by several clays all of which are mixed with washed silica sand (AFA fineness 63) except the Downer and Albany. These curves show the large differences which exist between the various clays and such an analysis may provide a means of identification, and evaluation.

41. Figures 30 and 31 show the result of blending two clays in an effort to obtain the most desirable properties of both. In Fig. 31, the middle curve shows the strength of 10 per cent fireclay in a 63.7 washed silica sand. When 2 per cent of the fireclay was replaced with 2 per cent of bentonite the strength shown in the top

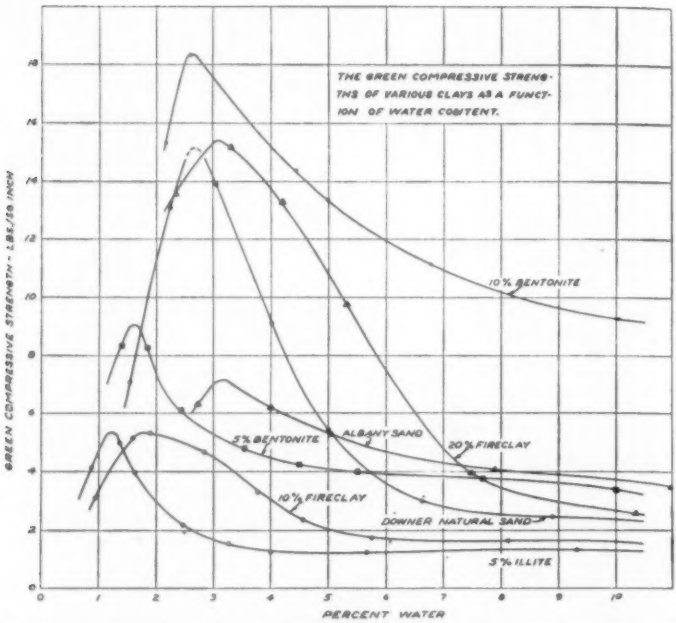


FIG. 29—GREEN COMPRESSIVE STRENGTHS OF VARIOUS CLAYS AS A FUNCTION OF THE WATER CONTENT.

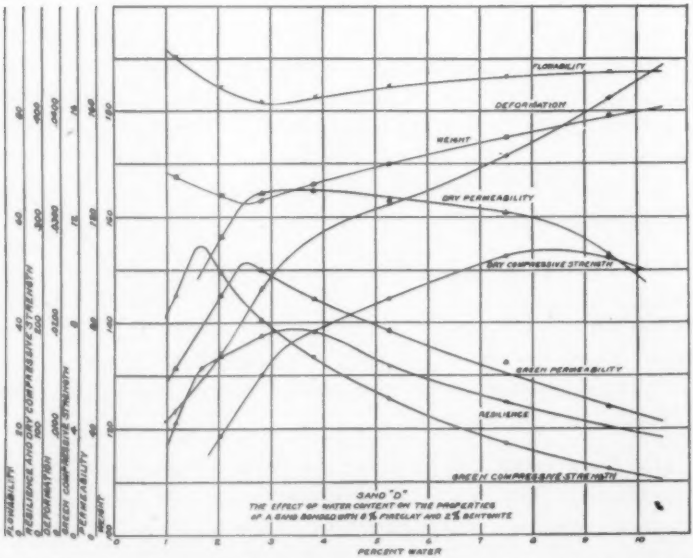


FIG. 30—EFFECT OF MOISTURE CONTENT ON THE PROPERTIES OF WASHED SILICA SAND *D*, TABLE 1, BONDED WITH A MIXTURE OF 8 PER CENT FIRE CLAY AND 2 PER CENT BENTONITE.

curve was produced. The strength of the mixed clay is much higher than the sum of the strengths of the two components. The sand with the mixed clay absorbed water somewhat more readily than the straight bentonite mix.

EFFECT OF CEREAL BINDER

42. The use of cereal binders in molding sands has been the subject of much discussion. Most steel foundries use them, claiming that they produce beneficial results. Usually the effect, however, is not stated in terms of the measurable physical properties of the sand but is described as an improvement in texture or in the behavior of the sand while it is being molded.

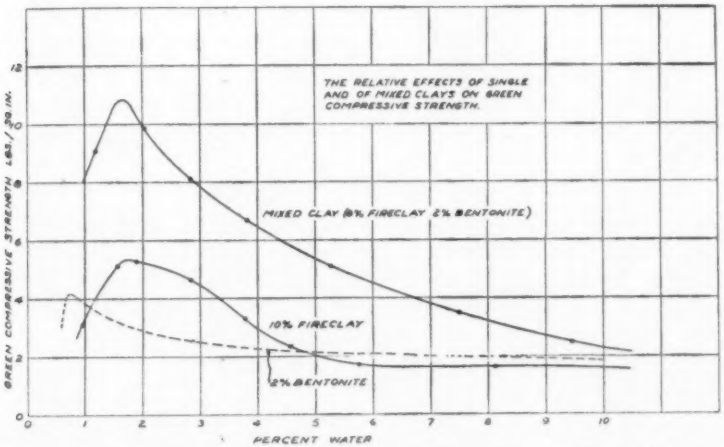


FIG. 31—THE RELATIVE EFFECTS OF SINGLE AND OF MIXED CLAYS ON GREEN COMPRESSION STRENGTH.

43. To study this subject, six mixes were made up containing 0, ½, 1, 2, 3, and 5 per cent of a popular and widely used cereal binder. The remainder consisted of 5 per cent bentonite and enough sand (63.7 A.F.A. fineness) to make 100 per cent. These materials were mulled dry for one minute, sufficient water was then added to produce 5 per cent moisture, and the mixture was mulled for 5 min. more. The six mixes were tested in the usual way and the results are shown in Fig. 32. The mix without cereal had good permeability and excellent flowability but felt wet and adhered slightly to the hands. The addition of small percentages of cereal caused the sand to feel much drier. At 1 per cent or more of cereal, the sand did not stick to the hands and seemed stronger.

At cereal contents of 2 to 3 per cent, a pronounced change in texture was observed. The individual coated grains united to form small pellets from $\frac{1}{8}$ to $\frac{1}{4}$-in. in diameter. At 5 per cent, this was still more pronounced. The sand felt dry, although the water content was held approximately constant. The agglomerated grains were quite large and the texture of the sand was so stiff that the sand could not be packed tightly. While the individual pellets were very compact, large interstices remained between them even after the sand was rammed hard.

Green Compressive Strength

44. Fig. 32 shows the effect of the cereal binder on the physical properties of the sand mixtures. Green compressive strength rises rapidly from 4.4 lb. per sq. in. without cereal to 8.1 lb. per sq. in. with 2.5 per cent cereal and then gradually decreases. At first glance it would appear that the cereal binder had approximately doubled the strength. There is another factor which must be considered, however. The strength produced in a

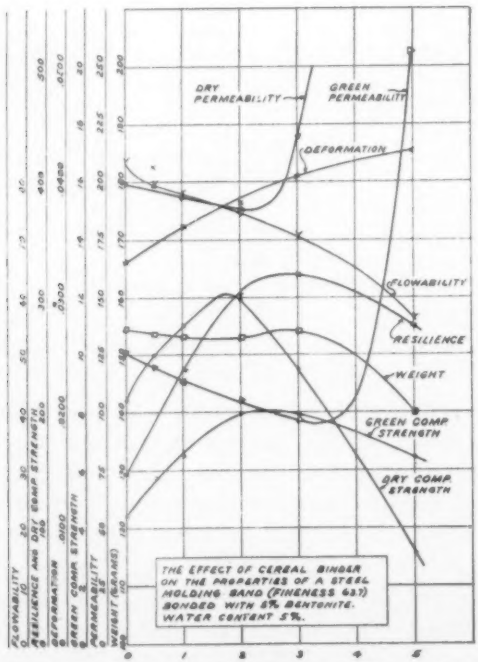


FIG. 32—EFFECT OF A CEREAL BINDER ON THE PROPERTIES OF WASHED SILICA SAND *D* BONDED WITH 5 PER CENT BENTONITE.

sand mixture by bentonite varies rapidly with the water content. When the cereal binder was introduced it absorbed part of the water in the mix and left a smaller amount of water available to the bentonite. With less water, the bentonite will produce a higher strength. This is shown very well in Fig. 9. From this, it is seen that in a mix containing the same grade of sand and the same amount of bentonite, a strength of 4.1 lb. per sq. in. was obtained at a water content of 5.0 per cent. When the moisture was reduced, however, the strength increased. This amounts to 8.2 lb. per sq. in. at 1.85 per cent water. This indicates that the increase in strength is not entirely due to the strength of the cereal binder but partly to its effect on the amount of water available to the bentonite. (Another mix discussed below shows the strength of the same cereal binder used without clay.)

45. It is debatable whether a mix should contain sand and clay with cereal binder and a moderately high moisture or whether the cereal should be omitted and less water used. Each method has its advantages. The omission of the cereal permits the reduction of water but forms a mix which may cause trouble in certain types of work by becoming dry and friable on the surface. Using the cereal, forms a mix which does not appear to dry out on the surface so rapidly. This mix will feel about as dry as the one without cereal but will contain more water.

Green and Dry Permeability

46. Most of the other properties in this series show the effect of the agglomeration of the grains at high cereal contents. Green permeability decreases from 150 to a minimum of 138 at 3 per cent cereal. Further additions of cereal cause a remarkable increase in permeability, due to the voids produced between the pellets. Dry permeability was very similar, starting at 179, decreasing to a minimum of 175 at 2 per cent cereal and then increasing very rapidly. At 5 per cent cereal, the dry permeability was so high that it could not be measured because the mercury leaked through the voids in the specimen.

Dry Compressive Strength

47. The dry compressive strength was 210 lb. per sq. in. without cereal, increasing to about 300 lb. per sq. in. at 2 per cent cereal and then decreasing rapidly with further additions. This corresponds roughly with the maximum in green strength. Unlike

green strength, this increase must be attributed solely to the cereal binder.

Flowability

48. Flowability, which was excellent without cereal (83 per cent), decreased rapidly when increasing amounts were added. At 5 per cent cereal, the flowability was reduced to 56 per cent.

Density

49. The weight of the specimen was nearly constant from 0 to 3 per cent cereal, and then the weight decreased rapidly. This indicates in another way the formation of voids in the specimen.

Specimen Deformation

50. For the specimens containing cereal binder, the fracture in compression in the green state was unusual in that it was very slow. As the load was applied, the test cylinder held its cylindrical shape for a time and then began to deform slowly into a barrel shape, indicating a high deformation before the final fracture. The deformation attachment, however, did not show this result as the needle usually started to recede before the final fracture occurred. The specimen was barrel-shaped after fracture with both ends pushed in ⅛-in. or more. Longitudinal cracks, due to circumferential tension, developed but the specimen did not break apart as is usually the case.

Effects of Cereal Binder Alone

51. Since the use of cereal binder with clay exhibited such interesting characteristics, it was thought desirable to investigate the effects of the material in mixes without clay to determine the strength which could be attributed directly to the cereal binder. A series of mixes was made up containing the same sand (fineness 63.7) and 2 per cent of the cereal. Water contents were varied from 0.66 per cent to 9.77 per cent and no clay was used. Figure 33 shows the physical properties of the series.

52. At water contents less than 2 per cent, the mixture was so weak that it was impossible to measure green compressive strength. The sand would not hold together. As the water was increased, the strength increased gradually to 1.47 lb. per sq. in. at 7 per cent water and then decreased. This strength is small compared with that of most clays. Another interesting point is that the shape of the green compressive strength curve is different

from that of all other series discussed in that the high strength, which occurs at low moisture contents with clay present, is missing when only cereal binder is used. Green permeability rises rapidly with increase in water, reaches a maximum of 245 at about 1 per cent water and decreases rapidly at first and then more gradually. The weight of the specimen, as would be expected, is just the inverse of the permeability curve. Dry permeability is approximately proportional to green permeability in the range tested.

53. The dry compressive strength increases rapidly to a maximum of about 800 lb. per sq. in. at 6.5 per cent water and then decreases. This is far in excess of the dry strength produced by a comparable mix containing clay. The dry compressive strength curve, from Fig. 32, for example, shows a dry strength of 298 lb. per sq. in. for a mix containing 5 per cent bentonite, 5 per cent water, and 2 per cent cereal. From Fig. 33, it is seen that the same material without clay has a strength of 700 lb. per sq. in.

SUMMARY

54. Cumulative curves drawn from sieve analyses are a convenient and effective method of showing the characteristics of

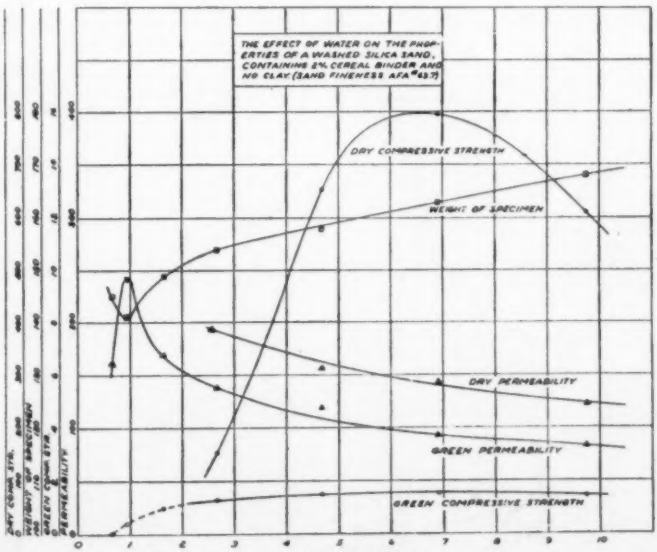

THE EFFECT OF WATER ON THE PROPERTIES OF A WASHED SILICA SAND, CONTAINING 2% CEREAL BINDER AND NO CLAY (SAND FINENESS AFA 45.7)
DRY COMPRESSIVE STRENGTH
WEIGHT OF SPECIMEN
DRY PERMEABILITY
GREEN PERMEABILITY
GREEN COMPRESSIVE STRENGTH
PERCENT WATER

FIG. 33—EFFECT OF MOISTURE CONTENT ON WASHED SILICA SAND *D* CONTAINING 2 PER CENT CEREAL BINDER AND NO CLAY.

Table 2

TEST RESULTS

No. of Mix	Type of Clay	Amount of Clay	Size of Sand	Amount of Sand	Other Bonding Material	Water Content	Weight of Specimen	Green Permeability	Green Compressive Strength	Deformation	Resilience	Flowability	Dry Permeability	Dry Compressive Strength
7A	Bentonite	5%	31.0	95%	None	1.35	169.0	417	9.28	0.0107	99.3	87.4	—	—
7B	"	"	"	"	"	1.90	165.0	467	11.31	0.0191	216.0	80.8	550	82.0
7C	"	"	"	"	"	2.53	165.0	440	8.23	0.0265	218.0	80.2	550	111.5
7D	"	"	"	"	"	3.51	168.0	370	6.70	0.0328	220.0	81.8	550	126.1
7E	"	"	"	"	"	4.50	171.0	326	5.90	0.0391	231.0	82.5	550	130.4
7F	"	"	"	"	"	5.55	173.5	287	5.11	0.0388	198.0	84.3	550	135.5
7G	"	"	"	"	"	7.45	177.5	224	4.39	0.0424	186.0	85.9	550	145.1
7H	"	"	"	"	"	10.30	184.5	126	4.00	0.0391	156.0	87.0	525	169.0
8A	Bentonite	5%	40.6	95%	None	1.23	165.0	268	8.59	0.0115	98.8	89.7	—	—
8B	"	"	"	"	"	1.79	161.0	313	12.27	0.0124	152.0	85.7	376	66.1
8C	"	"	"	"	"	2.36	159.0	341	8.37	0.0196	164.0	81.7	467	106.5
8D	"	"	"	"	"	3.45	162.0	317	6.16	0.0285	175.5	83.1	467	127.4
8E	"	"	"	"	"	4.30	164.0	287	5.20	0.0328	170.6	84.0	458	132.6
8F	"	"	"	"	"	5.41	166.5	253	4.54	0.0348	158.0	85.1	467	138.8
8G	"	"	"	"	"	7.41	171.0	190	3.89	0.0367	142.8	86.3	440	168.5
8H	"	"	"	"	"	9.64	175.0	135	3.55	0.0372	132.0	87.0	440	177.5
9A	Bentonite	5%	52.4	95%	None	1.17	156.5	153	6.16	0.0096	59.1	90.9	—	—
9B	"	"	"	"	"	1.70	152.5	203	9.25	0.0138	127.7	86.1	226	51.6
9C	"	"	"	"	"	2.48	151.5	212	6.38	0.0210	134.0	82.5	264	110.2
9D	"	"	"	"	"	3.46	153.5	207	4.92	0.0280	137.8	83.2	275	139.5
9E	"	"	"	"	"	4.44	156.0	186	4.47	0.0314	140.3	83.8	275	152.1
9F	"	"	"	"	"	5.29	157.0	178	3.94	0.0324	127.8	84.2	279	172.9
9G	"	"	"	"	"	7.68	161.0	142	3.29	0.0360	118.4	85.3	283	200.2
10A	Bentonite	5%	63.7	95%	None	1.38	152.5	120	8.31	0.0116	96.4	87.2	—	—
10B	"	"	"	"	"	1.83	149.5	145	8.23	0.0141	117.3	83.1	173	69.5
10C	"	"	"	"	"	2.44	149.5	146	6.03	0.0195	117.5	80.6	188	109.4
10D	"	"	"	"	"	3.53	151.5	141	4.80	0.0264	126.7	82.0	193	154.8
10E	"	"	"	"	"	4.49	153.2	133	4.21	0.0304	128.0	82.4	195	180.7
10F	"	"	"	"	"	5.51	155.0	123	3.96	0.0314	124.3	84.0	198	202.6
10G	"	"	"	"	"	7.69	159.0	101	3.72	0.0326	121.2	84.2	198	235.9
10H	"	"	"	"	"	10.01	163.0	82	3.33	0.0328	109.2	84.8	193	265.1
11A	Bentonite	5%	79.8	95%	None	1.46	158.0	81	12.11	0.0111	134.6	89.1	—	—
11B	"	"	"	"	"	1.92	153.5	94	11.35	0.0154	174.7	83.6	120	—
11C	"	"	"	"	"	2.41	153.5	99	7.90	0.0218	172.3	81.4	129	—
11D	"	"	"	"	"	3.63	156.0	90	5.85	0.0293	171.4	82.7	131	—
11E	"	"	"	"	"	4.48	157.5	88	5.03	0.0344	173.0	83.5	137	—
11F	"	"	"	"	"	5.42	159.5	81	4.64	0.0340	157.7	83.7	134	—
11G	"	"	"	"	"	7.99	164.5	61	4.15	0.0409	169.7	85.0	126	—
11H	"	"	"	"	"	10.25	169.0	46	3.87	0.0385	149.0	85.6	116	—
12A	Bentonite	5%	120.3	95%	None	1.53	156.5	36	12.36	0.0105	129.7	88.3	40	39.8
12B	"	"	"	"	"	2.02	153.0	40	13.62	0.0122	166.2	82.8	48	83.7
12C	"	"	"	"	"	2.57	151.0	44	10.53	0.0171	180.2	80.5	57	142.6
12D	"	"	"	"	"	3.50	153.5	44	7.58	0.0253	192.0	80.3	60	195.8
12E	"	"	"	"	"	4.48	155.5	42	6.52	0.0302	197.0	80.1	60	213.9
12F	"	"	"	"	"	5.60	157.5	40	5.88	0.0302	178.0	80.6	58	242.7
12G	"	"	"	"	"	7.66	162.0	34	5.20	0.0359	186.7	82.6	58	275.1
12H	"	"	"	"	"	10.00	166.0	29	4.84	0.0351	170.0	82.8	52	—
18A	Bentonite	5%	67.6	95%	None	1.18	160.5	70	9.20	0.0102	109.2	89.4	—	—
18B	"	"	"	"	"	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]
18C	"	"	"	"	"	9.39	165.0	57	4.25	0.0324	141.2	84.7	139	166.6
18F	"	"	"	"	"	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]
19A	Bentonite	5%	82.9	95%	None	1.22	171.0	39	10.65	0.0105	111.8	88.4	—	—
[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	1.59	166.5	46	14.74	0.0117	172.5	83.7	54	59.9
[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	[illegible]	76.9	89	103.1

18A	Bentonite	5%	67.6	95%	None	1.18	160.5	79	9.29	0.0102	129.3	89.4	—	—	
18B	"	"	"	"	"	9.59	165.0	57	4.28	0.0334	141.2	85.7	139	166.6	
18F	"	"	"	"	"										
19A	Bentonite	5%	82.9	95%	None	1.22	171.0	39	10.65	0.0105	111.8	88.4			
19B	"	"	"	"	"	1.59	166.5	46	14.74	0.0117	172.5	83.7	54	59.9	
19C	"	"	"	"	"	2.12	161.0	59	11.64	0.0161	187.5	76.9	80	103.1	
19D	"	"	"	"	"	2.94	164.5	55	8.36	0.0222	185.5	78.0	80	160.6	
19E	"	"	"	"	"	5.39	170.5	45	6.28	0.0276	173.3	78.7	80	204.7	
19F	"	"	"	"	"	10.02	180.5	29	5.93	0.0346	205.1	81.0	67	255.1	
25A	Fireclay	10%	63.7	90%	None	0.94	162.0	61	3.09	0.0134	41.4	91.5	—	—	
25B	"	"	"	"	"	1.58	161.0	71	5.14	0.0114	58.6	89.7	—	—	
25C	"	"	"	"	"	1.90	160.0	88	5.29	0.0118	62.4	87.4	96	13.5	
25D	"	"	"	"	"	2.84	159.0	103	4.65	0.0135	62.6	84.5	126	27.5	
25E	"	"	"	"	"	3.78	161.0	108	3.29	0.0190	62.5	83.8	139	43.1	
25F	"	"	"	"	"	4.56	162.0	108	2.38	0.0233	55.5	85.2	145	47.1	
25G	"	"	"	"	"	5.76	164.0	116	1.72	0.0236	40.6	86.7	156	53.5	
25H	"	"	"	"	"	8.12	170.0	93	1.64	0.0214	35.5	86.4	122	73.5	
26A	Fireclay	20%	63.7	80%	None	2.31	169.5	37	13.58	0.0100	135.8	83.1	44	32.8	
26B	"	"	"	"	"	3.29	169.5	50	15.16	0.0138	209.4	79.6	60	47.9	
26C	"	"	"	"	"	4.18	170.0	55	13.24	0.0223	295.5	77.8	74	65.8	
26D	"	"	"	"	"	5.31	172.0	58	9.75	0.0310	302.0	76.7	81	80.1	
26E	"	"	"	"	"	7.48	180.0	62	3.91	0.0384	150.0	81.3	102	90.3	
26F	"	"	"	"	"	10.35	189.0	40	2.59	0.0386	100.0	84.6	76	109.0	
27A	Downer	10.72%	52.1	89.28%	None	1.54	163.0	44	7.10	0.0123	87.3	89.2	—	—	
27B	"	"	"	"	"	2.22	161.5	66	13.05	0.0137	178.8	85.4	—	—	
27C	"	"	"	"	"	3.02	158.0	108	13.90	0.0133	185.0	79.4	126	24.8	
27D	"	"	"	"	"	4.00	159.0	111	9.07	0.0208	188.5	78.6	138	38.0	
27E	"	"	"	"	"	5.00	162.0	108	5.42	0.0261	141.3	79.5	142	43.3	
27F	"	"	"	"	"	6.64	168.0	94	3.00	0.0328	98.4	82.5	142	63.9	
27G	"	"	"	"	"	8.90	173.0	71	2.45	0.0443	108.3	83.8	128	89.0	
28A	Illite	5%	63.7	95%	None	0.88	163.0	89	4.13	0.0112	46.2	91.3	—	—	
28B	"	"	"	"	"	1.36	160.0	112	4.97	0.0132	65.6	88.7	—	—	
28C	"	"	"	"	"	1.63	159.0	129	3.93	0.0133	52.3	86.3	—	—	
28D	"	"	"	"	"	2.47	161.0	126	2.18	0.0220	48.0	86.5	151	42.5	
28E	"	"	"	"	"	3.27	162.0	129	1.53	0.0256	39.2	87.0	161	63.7	
28F	"	"	"	"	"	4.00	163.7	131	1.22	0.0300	38.1	88.4	167	66.1	
28G	"	"	"	"	"	5.67	167.0	114	1.22	0.0341	41.6	88.5	145	93.3	
28H	"	"	"	"	"	9.32	176.5	64	1.30	0.0373	48.5	88.8	94	134.0	
30A	Fireclay	8%	63.7	90%	Bentonite 2%	1.18	167.5	63	9.05	0.0116	105.0	90.2	—	—	
30B	"	"	"	"	"	2.04	164.0	90	9.88	0.0170	168.0	84.4	112	93.1	
30C	"	"	"	"	"	2.82	163.0	100	8.09	0.0231	186.8	81.6	129	150.0	
30D	"	"	"	"	"	3.82	166.0	89	6.70	0.0306	205.0	82.2	130	190.7	
30E	"	"	"	"	"	5.26	170.0	77	5.09	0.0312	158.8	84.8	126	223.0	
30F	"	"	"	"	"	7.50	175.0	65	4.36	0.0358	124.0	86.1	122	263.8	
30G	"	"	"	"	"	9.46	179.0	48	2.48	0.0413	102.0	87.1	105	260.4	
32A	Bentonite	5%	63.7	95.0%	None 5% (approx.)	154.5	126	4.42	0.0330	146.0	83.3	198	210.3		
32B	"	"	"	94.5	Cereal 0.5%	"	153.5	119	5.78	0.0412	238.0	79.4	198	250.0	
32C	"	"	"	94.0	" 1.0	"	153.0	113	6.54	0.0361	236.0	78.0	193	275.0	
32D	"	"	"	93.0	" 2.0	"	153.0	105	7.94	0.0381	302.0	74.7	188	298.0	
32E	"	"	"	92.0	" 3.0	"	154.0	97	7.91	0.0404	320.0	70.8	219	237.0	
32F	"	"	"	90.0	" 5.0	"	140.0	257	6.46	0.0425	275.0	56.2	Very high	77.7	
33A	None	—	63.7	98%	Cereal 2.0%	0.66	145.0	161	Very low						
33B	"	"	"	"	"	0.94	141.0	243	Low						
33C	"	"	"	"	"	1.63	149.0	167	"						
33D	"	"	"	"	"	2.67	154.0	138	1.27				193	153.1	
33E	"	"	"	"	"	4.68	158.0	119	1.44				156	651.0	
33F	"	"	"	"	"	6.90	163.0	93	1.47				142	795.0	
33G	"	"	"	"	"	9.74	168.0	84	1.41				122	609.0	

molding sands. They show at a glance the size, distribution, clay and silt contents and any unusual characteristics of the sand being studied.

55. Sand grain size changes the green compressive strength-water content curve in magnitude but not in shape. The sands of intermediate sizes (A.F.A. Fineness 50 to 60) are weaker than either the finer or the coarser sands. Dry compressive strength increases as the grain size decreases. The density of the rammed sand is much higher for the very coarse sands (A.F.A. Fineness 30-40) than for the medium and fine sands.

56. Sands of uniform grain size produce higher green and dry permeability and flowability than do non-uniform sands. The non-uniform sand produces higher strength in the green and dry states and higher density in the rammed material.

57. Since each of the clays tested produces a green compressive strength-water content curve which is characteristic of the clay, the shape and magnitude of this curve may provide a method of estimating the mineral constitution of a clay. Each clay would have to be tested at several water contents with a standard sand and compared with curves of relatively pure clay minerals also tested with the standard sand.

58. Small amounts of bentonite added to sands containing other clays give strengths higher than the sum of the strengths of the two clays tested separately.

59. Bentonite sands dry out on the surface more rapidly than fireclay bonded sands because the water from the backing sand does not diffuse into the surface layer as readily as it does in fireclay bonded sands.

60. Cereal binder (corn flour type) increased green and dry strength up to 2.5 per cent cereal, but reduced these values as further cereal was added. The increase in green strength may, in part, be due to the absorption of water by the cereal, allowing the bentonite to work at a more efficient clay to water ratio, since the cereal produces very little green strength when used alone. The use of over 2 per cent of cereal caused the sand to ball up into small pellets, which acted as individual grains. This produced abnormally high permeability, low flowability and soft spots in the sand.

61. Table 2 shows the results obtained in the series of tests performed during the investigation.

ACKNOWLEDGMENT

62. Grateful acknowledgment is made herein to the following members of the Naval Research Laboratory Staff: Dr. F. M. Walters, E. A. Rominski, and C. E. Jackson, for editing the manuscript, encouraging the work, and for helpful suggestions at various times. The authors are also greatly indebted to the Navy Department for sponsoring the work and for permission to publish the results.

Bibliography

1. Dietert, H. W., *"Commercial Application of Molding Sand Testing,"* TRANSACTIONS, American Foundrymen's Association, vol. 32, part 2, pp. 24-46 (1924).

2. Hansen, C. A., *"The Physical Properties of Foundry Sands,"* TRANSACTIONS, American Foundrymen's Association, vol. 32, part 2, pp. 57-97 (1924).

3. Adams, T. C., *"Permeability Testing of Molding Sands,"* TRANSACTIONS, American Foundrymen's Association, vol. 32, part 2, pp. 114-167 (1924).

4. Grim, R. E., and Rowland, R. A., *"Relationship Between the Physical and Mineralogical Characteristics of Bonding Clays,"* TRANSACTIONS, American Foundrymen's Association, vol. 48, pp. 211, 224 (1940).

5. *"Report of the Committee on Grading,"* TRANSACTIONS, American Foundrymen's Association, vol. 38, pp. 510, 515 (1930).

6. Krumbein, W. C., *"The Size Frequency Distribution of Sediments,"* JOURNAL OF SEDIMENTARY PETROLOGY, vol. 4, no. 2, pp. 65-77.

7. Dietert, H. W., Woodliff, E. E., and Schuck, J. A., *"A Study of Steel Molding Sands,"* TRANSACTIONS, American Foundrymen's Association, vol. 46, pp. 257-279 (1938).

8. Sheehan, J. J., *"Core-shop Control,"* FOUNDRY TRADE JOURNAL, June 15, 1939, pp. 516-519.

9. Briggs, C. W., and Morey, R. E., *"Synthetically Bonded Steel Molding Sands,"* TRANSACTIONS, American Foundrymen's Association, vol. 47, pp. 653-724 (1939).

10. *"Testing and Grading Foundry Sands and Clays—Standards and Tentative Standards,"* 1938 Edition. Published by the American Foundrymen's Association.

DISCUSSION

Presiding: J. B. CAINE, Sawbrook Steel Casting Co., Lockland, Ohio.

N. J. DUNBECK[1]: I would like to know whether or not the speaker has been able to correlate his findings on flowability, as determined in the standard test, with the presence of swells in castings, which is probably the one defect which is common, or assumed to be so anyway?

MR. MOREY: The setup we have does not allow a great number of molds to be made so that we do not have as good a chance to study a thing of that sort as you would in a commercial foundry. Our foundry is small. But we do find that unless we keep the hardness up pretty well, we get swells.

MR. DUNBECK: Would you find any correlation between the hardness of a specimen and the flowability?

MR. MOREY: You probably would find some. That is, low flowability would tend to have spots of low hardness in the mold.

MR. DUNBECK: Low flowability and low hardness together?

MR. MOREY: Low flowability would produce low hardness in spots.

MEMBER: We have been doing some work on the density and flowability of sands at the University of Michigan. We have used quite a few mixtures of almost the same composition as Mr. Morey has used. I suppose Mr. Morey has found that, if the clay or bentonite content is increased 5 per cent, it is impossible, with that method, to get a specimen that is uniform in density throughout. It is very hard on top and very spongy at the bottom. I do not think that a specimen of that kind is representative for test. If the specimens are constantly dense, as is done in England, we will get almost a straight line function instead of these variations.

MR. MOREY: With regard to the method used in England, that is, ramming the sands to constant density, the sands will have to be rammed much harder in the high water content range than in the very low water content range and that, of course, will influence the various properties.

MEMBER: I do not try to defend the British specimen. There might be many things against ramming to constant density, but I think the method is a little too sensitive to really give results that can be directly transplanted to molding in a foundry.

MR. MOREY: I think myself that a lot more work needs to be done on flowability.

E. PRAGOFF, JR.[2]: I would like to compliment the authors on the

[1] Vice President, Eastern Clay Products, Inc., Elfort, Ohio.
[2] Sales Supervisor, Hercules Powder Co., Wilmington, Del.

type of work they are doing. In spite of the fact that we know it is difficult always to incorporate our laboratory results in our results on regular castings, I still think they have given us a very nice tool to work with. They are leading us to think more in terms of studying the effect of variables over a wide enough range that we can intelligently prepare better molds with the object of making better castings.

There was one question I had in my mind. All your samples are prepared the night before, or for a 24-hour period prior to testing. Why?

MR. MOREY: We find that, to get comparable results, to check them day after day, that we must follow a very rigid procedure in the preparation of the samples. The amounts are weighed to grams, and they are mulled dry in an 18-in. laboratory muller for one minute and then the water is added and they are mulled for 5 min. They are dumped into 2-qt. Mason jars, sealed and allowed to temper for 24 hours. Then they are removed from the Mason jars and passed as rapidly as possible through a ¼-in. sieve, which tends to homogenize the sand. That is done at the same time for all of the 8 to 10 mixes in a series. Then the mixes are tested as rapidly as possible one after another so that the conditions of temperature, humidity, etc., are similar for all the batches in that series.

MR. PRAGOFF: You selected the 5-min. mixing period as that would be a fairly good period for all these mixes you prepared, whether they are one or 10 per cent bentonite or one or 10 per cent moisture, or did you select that period which you decided was the optimum?

MR. MOREY: We know, that for 2 or 3 per cent bentonite, we do not need that much time, and for 10 per cent bentonite, that is the very heavy mixes, we would probably be better off with more time than that. That was the commonly used average.

MR. PRAGOFF: The question of mulling time is something we all ought to think more about in the preparation of facing sand. I have seen good facings made in a short mulling time. I have also seen some not completely mulled and not ready to use. We have expensive equipment, we make good sand, we pay money for binders, and we do not get all that we have paid for because we do not mix properly. In some cases, the batch might be over-mulled.

MR. MOREY: There is a compromise between getting the best mulling and getting the most sand out.

MR. PRAGOFF: One other point about that question of making the sand samples the night before. I wonder if any steel foundries make up their facing sand the day before it is used? I do not know of any. I wonder whether your work would be an indication that sand is better if it is prepared a day in advance? Do you think that would be true?

MR. MOREY: In our case, on the floor we make up the facing the same day that we use it. We do not follow that practice in the test work.

MR. PRAGOFF: I would propose that in the test work you follow the procedure that the foundryman is going to follow.

MR. MOREY: I have compared the two methods and found very little difference in them, for the sands I have used.

MR. PRAGOFF: If you made up a batch on Friday night and had it left over until Monday, it would be all right as long as it was sealed up?

MR. MOREY: Yes. In this case it is very important to sieve the sand before testing.

MR. DUNBECK: Mr. Morey mentioned that he got some interesting green strengths when he used a combination of fire clay and bentonite. That is undoubtedly true. We think even more interesting figures are given when we look at the dry strength of combinations, since the combination of the two materials will give peak dry strength very much higher than either material used separately. We know that if we put in enough water with either fire clay or bentonite, we can set up very high dry strength, but it also costs money to get that water out in drying molds. In using the combinations which give a very high dry strength of what we might call normal or comparatively low water content, we obtain real economy. We also have the attractive feature that the addition of a low cost material like fire clay to a high cost material like bentonite, gives a higher dry strength than the high cost material when used straight.

MR. PRAGOFF: Why do you get a higher dry strength with that combination?

MR. DUNBECK: I have no idea. That combination will set up very high peaks in hot strengths, although the combination of other types of clays will go in a straight line from the low to the high strength clay. I do not know why it happens.

MEMBER: I have been in shops where the straight bentonite mixture has been changed to a 60-40 bentonite clay-mixture to improve shake-out conditions. This would be contrary to the statement just made, and I wondered why and whether there were figures to prove that.

MR. DUNBECK: They are absolutely wrong if they do that. There would be more lumps in the shake-out with that combination.

Report of Steel Division Committee on Radiography — 1941

To the Steel Division, American Foundrymen's Association:

Your Radiographic Committee in January of this year sent out a questionnaire on "Radiography in the Steel Foundry." This questionnaire was mailed by the Association to 189 steel foundries in the United States and one steel foundry in Canada. The questionnaire was modeled after the 1938 questionnaire with a few additional questions included. It was stated in the questionnaire that if the foundry organization answering the questionnaire wished to remain anonymous, it was at liberty to do so.

The committee received 79 replies from the 190 questionnaires sent out, a return of about 41 per cent. The committee feels that this is an excellent return in view of the fact that in 1938 replies were received from only 20 per cent.

Of the 79 that replied to the questionnaire 42 stated that they had used radiography. This means that approximately 58 per cent of those that made returns had made some study of the subject. The committee is inclined to believe, however, that practically all those that have experienced the use of radiography made returns. If this condition is true, then 22 per cent of all the steel casting producers had had actual radiographic experience.

The committee has not discounted the replies from 37 organizations which had not used radiography. In fact, some very interesting information was made available by their replies, for they not only show the trend that they desire radiography to take, but they also show a live interest in the subject.

Questionnaire Replies

In the following pages, the committee reports in detail the questions as submitted and the answers received. Comparison be-

Note: This paper was presented before a steel session at the 45th Annual A.F.A. Convention, New York City, N. Y., May 15, 1941.

tween the 1941 and the 1938 answers are made. The committee also has taken the liberty to comment on these answers.

1. *Has your organization used X-ray or Gamma ray nondestructive testing in the manufacture or inspection of steel castings?*

	1941	*1938*
Yes	42	19
No	37	30

2. *If the answer to (1) is "yes," please check one of the following:*

		1941	*1938*
(a)	Rarely, for special examination	12	10
(b)	Occasionally	9	4
(c)	Regular procedure	8	5
(d)	Used to improve your own foundry technique	22	..
(e)	Compulsory for inspection	18	..

(Several foundries checked more than one answer.)
It is interesting to note that the majority are using radiography for the improvement of foundry technique.

3. *Which method are you using?*

(a)	X-rays	9
(b)	Gamma rays	29
(c)	Both	4

4. *Will you take orders for castings that are subject to radiography?*

Yes	45
No	18
Failed to answer	16

An interesting point is that 9 returns from the group that have not used radiography answered "yes." Two foundries that have used radiography said they would not take orders that are subject to radiography. No reasons were given. Presumably these foundries have not had sufficient experience with radiography as yet.

5a. *Have your customers used X-ray or Gamma ray radiography for inspection?*

	1941	*1938*
Yes	37	22
No	19	17
Failed to answer	23	10

This shows an increase in customers going to radiographic inspection.

5b. *Have the results of this inspection been mutually satisfactory?*

	1941	*1938*
Yes	36	13
No	14	9
Failed to answer	6	0

The replies show that with greater understanding of the subject a more tolerant viewpoint is effected.

More information may be obtained if the answers are broken down into two groups, those that have used radiography and those that have not, as shown by the answer to question 1. These groups will be termed "Yes Radiography" and "No Radiography." Answers to question 5b break down as follows:

1941 Group	*Yes*	*No*
"Yes" Radiography	29	6
"No" Radiography	7	8

Foundries that use radiography apparently are more able to see the consumers point of view.

One producer in commenting on his answer to 5b stated that he had answered in the affirmative though in a few cases the results were not satisfactory.

6. *Have you manufactured castings subject to Navy Radiographic Standards?*

Yes	17
No	17
Failed to answer	45

This answer is better understood by examining the break down since the group that has used radiography would be the ones most concerned.

Group	*Yes*	*No*	*Failed to answer*
"Yes" Radiography	15	17	10
"No" Radiography	2	0	35

Approximately 50 per cent of those who have used radiography have manufactured castings subject to Navy Radiographic Standards.

7. *Do you believe that the gamma ray standards of the Navy Department for steel castings for steam pressure service are*:

(a)	Too lenient	1
(b)	Fair	14
(c)	Unnecessarily strict?	4
(d)	Comprehensive enough?	..
(e)	Adhered to, if used?.....	1
(f)	In need of more clarification?.....	15

There is apparently an equal division between those producers that think the standards are fair and those that believe more clarification is needed.

A producer in answering question 7 stated: "Standards are too lenient for small castings under ¾" wall thickness." Another producer had the following to say concerning his answer to question 7: "We answered 'Fair' providing, of course, that the Standards are really applied, which has not always been our experience. We have had cases where defects which were acceptable according to Navy Standards had to be chipped out and welded as the inspector only would submit perfect films to the Bureau in Washington to avoid questions from there. We also checked 'in need of more clarification' as the defects shown in Standards are not tied in with wall thickness of the section which is shown in the standard films."

8a. *Do you believe that the Navy Standards best serve the purpose for which they were devised?*

Yes	20
No	9
Failed to answer	50

8b. *Would you welcome the adoption of these standards by the American Society for Testing Materials?*

Yes	18
No	18
Failed to answer	43

This is broken down as follows:

Group	*Yes*	*No*	*Failed to answer*
"Yes" Radiography	16	9	17
"No" Radiography	2	9	26

The break down results show that those that have used radiography welcome adoption by approximately 2-1 even though the total answers are equally divided.

8c. *Would you be willing to use those same standards for services other than steam pressure?*

Yes	17
No	18
Failed to answer	44

Here again the break down into groups shows more information:

Group	*Yes*	*No*
"Yes" Radiography	15	10
"No" Radiography	2	8

One producer in commenting on question 8 stated: "It is our feeling that the present Navy Standards need clarification as to details and before any specification making body similar to the A.S.T.M. promulgates similar standards, the details should be very carefully worked out between producer and consumer." Several other producers in answer to 8a stated they agreed if the standards were clarified.

In reply to question 8c, one producer qualified his answer by stating that he would be willing to use these standards for services other than steam pressure "if the castings were small, but not for large castings for structural purposes." Another producer agreed to their use "in highly stressed sections; otherwise, no."

Apparently those that understand Radiography are tentatively willing to extend the use of the Navy Standards to other types of castings.

9. *In inspection and acceptance tests requiring radiography, should reference be made as to what areas or sections of the castings are to be radiographed?*

Yes	57
No	1
Failed to answer	21

Foundrymen are overwhelmingly in favor of this proposition. Consumers please note.

10. *Who should have authority for casting rejections?*

		1941	*1938*
(a)	Producer?	3	8
(b)	Consumer?	2	1
(c)	Consumer — with radiographic standards as a guide for rejection?	9	..*
(d)	Umpire (radiographic expert)?	3	9
(e)	By mutual agreement of producer and consumer?	41	33
	Failed to answer	21	

* Not in 1938 questionnaire.

As in 1938, the outstanding feature is that the steel casting manufacturers believe rejection can be best settled by mutual agreement between the producer and the consumer.

11. *Do you hesitate to use radiography?*

	1941	*1938*
Yes	31	31
No	36	11
Failed to answer	12	7

The break down of this answer shows:

	1941		*1938*	
Group	*Yes*	*No*	*Yes*	*No*
"Yes" Radiography ..	9	30	9	6
"No" Radiography ..	22	6	22	2

In the totals it may be noticed that there is considerable change in opinion in 3 years. In 1938, three-fourths of those replying hesitated to use radiographic examination. At the present time, more than 50 per cent do not hesitate to use radiography. As can be seen from the break down studies, this is due to the fact that radiography is better understood and foundrymen are more familiar with what is required.

12. *If the answer to (11) is "yes," please check one or more of the following reasons:*

		1941	*1938*
(a)	Unfamiliarity with procedures	18	10
(b)	Fear of incorrect interpretation of results	14	21
(c)	Fear of injury to personnel	4	7
(d)	Cost of radiographic equipment	18	27
(e)	Cost of radiographic examination	17	21
(f)	Foundry not equipped for high quality work	5	..

The answers are not much different from those in 1938. The break down results present a little more information:

	a	*b*	*c*	*d*	*e*	*f*
"Yes" radiography 1941....	2	6	0	4	5	0
"No" radiography 1941....	16	8	4	14	12	5
"Yes" radiography 1938....	1	9	2	6	5	..
"No" radiography 1938....	9	12	5	21	16	..

Those that use radiography today consider that the "fear of incorrect interpretation" is of the greatest importance whereas "cost of equipment" and "unfamiliarity with procedures" stand

out as the main reasons for the 1941 group that have not used radiography. It is apparent from this study that if these organizations would take up radiography, they would no longer consider that cost and unfamiliarity with procedures are so important.

13. *Has radiography been instrumental in causing beneficial improvement in manufacturing methods?*

	1941	*1938*
Yes	35	12
No	5	4
Failed to answer	2	3

It may be noted that a high percentage of those using radiography believe that it has produced beneficial results. The percentage has increased materially since 1938.

14a. *Would you be willing to use radiographic examination in production work to improve casting technique?*

	1941	*1938*
Yes	52	37
No	9	6
Failed to answer	18	6

14b. *If the answer to (a) is "no," are you fearful that because of such equipment the customer will demand more than you are capable of producing?*

	1941	*1938*
Yes	5	3
No	4	3
Failed to answer	0	0

14c. *Would you be willing to use radiography if your customers did not know you had such equipment?*

	1941	*1938*
Yes	26	18
No	5	2
Failed to answer	48	22

The answers to question 14 were similar to those in 1938. It is interesting to note that so many producers are willing to use radiographic examination in connection with studies in their casting technique. It is evident that they believe that radiographic inspection is of benefit to the foundryman. Only 9 producers felt that they would not use it and 5 of these were afraid that the presence

of radiographic equipment might influence their customers to make unjustifiable demands.

Of the 52 that answered "yes" to question 14a, 17 of them were producers that had not used radiography. It is evident, by comparing the 1938 and 1941 answers, that these 17, who have not used radiography in 1941, look favorably upon it and will probably be using radiography before another three years have passed.

15. *Where radiographic examination is used in acceptance tests, should this inspection be covered by radiographic standards?*

	1941	*1938*
Yes	52	33
No	2	5
Failed to answer	25	11

A very large majority definitely feel that radiographic inspection should be covered by specifications.

A producer commented on this question as follows: "The acceptance or rejection of material on radiographic examination should be covered by radiographic standards which are comprehensive, well clarified and acceptable to both producer and consumer. As it is impossible to obtain absolute perfection in commercial steel castings, any standard used should be graded to properly take care of the requirements for that particular application without unnecessarily adding to the cost burden."

16. *Would you welcome the preparation by the American Society for Testing Materials of standards for X-ray radiography for steel castings?*

Yes	51
No	5
Failed to answer	23

It is very evident that the steel casting producers would welcome standards prepared by the A.S.T.M.

17. *Do you think that radiographic specifications should specify details of radiographic technique?*

	1941	*1938*
Yes	37	25
No	13	9
Failed to answer	29	15

Of those that have used radiography, 12 feel that the technique

of radiography should not be specified but should be left to the radiographer to carry out as he deems best.

A producer commenting on his answer stated: "Specifications should specify details insofar as they can be standardized." Another producer stated, "We believe that the word 'technique' could cover many phases of handling and inspection. Radiographic specifications include all technique that could apply *in general,* such specified technique to be subject to modification and amplification as required for each individual case of inspection."

18a. *Please give some indication of your monthly tonnage output.*

Tons - range	*1941*	*1938*
0 to 99	5	10
100 to 499	39	16
500 to 999	16	13
Over 1000	13	10

18b. *Size of largest regularly produced castings:*

	1941	*1938*
½ to 100 lb.	12	6
100 lb. to 1 ton	21	13
1 ton to 5 tons	17	14
5 tons to 10 tons	12	6
10 tons to 25 tons	2	6
Above 25 tons	7	4

18c. *Types of castings produced to radiographic requirements.*

Ordnance	15	Airplane	7
Steam	15	Hydroelectric	5
Ship	13	Heat and Corrosion	4
Refinery	10	Automobile	2

19a. *Did you submit an answer to the questionnaire sent out by the committee in 1939?*

Yes	36
No	19

19b. *Is your attitude toward radiography more favorable today than in 1938?*

Yes	39
No	8
Failed to answer	32

20. *What comments do you desire to make?*

The following comments were received:

1. "We feel that standards of acceptance are necessary. Present Navy standards are not used at our plant by mutual agreement because they are too lenient for small castings of under ¾-in. wall section. However, lack of some standards has been cause for considerable disagreement, not only between producer and consumer but also between inspectors of the same consumer (U. S. Navy)."

2. "We believe in X-ray or Gamma ray properly done and interpreted."

3. "Physical examination technique covering operator's condition should be more positive than by present blood examination."

4. "Development of differing methods is being watched, and when demand necessitates it, one or more methods doubtless will be adopted."

5. "We believe the foundryman should be given a drawing of any job that is subject to non-destructive test, marked plainly exactly where tests are to be made. Then after careful study the management can decide if it is advisable to undertake the job without further understanding and discussion with the consumer."

FUTURE PLANS OF THE COMMITTEE

The committee believes that the above report should be received by members of the Steel Casting Industry as an indication of the second period or the development stage of radiography in the steel foundry. It is the hope of the committee that another questionnaire, embodying perhaps the same set of questions, may be sent out after a period of 3 to 5 years. In this manner, the committee may be helpful to steel casting manufacturers by showing them the trend of the times.

RADIOGRAPHY DURING 1940-1941

Your Radiographic Committee during the past year (May

1940 to May 1941) has followed radiographic developments carefully.

Committee E-7 on Radiographic Testing of A.S.T.M. formed a special sub-committee during the year to study the possible preparation of X-ray standards for steel castings. A search for those organizations that could furnish material for standards was instigated.

The gamma ray standards of the Navy Department, which were approved by E-7, were returned to this committee by E-10, the Committee on Standards, with the suggestion that they be clarified.

A committee was formed by E-7 to revise and materially extend E15-39T, Recommended Practice for Radiographic Testing of Metal Castings. So far work has not been started.

Committee E-7 sponsored two technical papers at the 1940 Annual Meeting of the A.S.T.M.

A paper by H. E. Seeman[1] pointed out that to secure the best X-ray radiograph, it is often necessary to mask irregular objects or to make separate exposures for the thick and thin parts. The need for such procedures can be lessened, or avoided entirely, by the use of a filter in the X-ray beam to diminish subject contrast and to decrease the amount of secondary radiation reaching the image from portions of the cassette not covered by the object. Seeman's experiments show that considerable improvement in radiographic quality can be made by placing the filter as near the X-ray tube as possible. Several advantages are listed for this type of technique and exposure data are listed for several representative subjects.

G. E. Doan and W. H. Sharp[2] presented a paper which continued the subject of precision radiography. The authors used gamma rays to study cracks of variable depths and widths in plates of varying thicknesses. The results show that there are definite limits beyond which internal flaws will not be revealed by radiographic methods. They presented limiting values showing the relation of crack width to crack depth.

During the year, articles on radiography other than those sponsored by the A.S.T.M. appeared in the industrial press.

[1] Superior numbers refer to references at end of report.

An article by R. G. Tobey and H. E. Seeman[3] pointed out that lead foil, as frequently used for screens in the radiography of materials, is awkward to handle because of its excessive pliability. Mounting the foil on sheet aluminum or on cardboard gives a rigidity which facilitates handling and does not interfere with the radiographic properties. Preparation of such screens is described in detail.

Recently an X-ray equipment rated for a maximum potential of one million volts was installed. Studies on its penetration power of steel and its ability to detect defects have been reported[4]. A chart shows that 9 in. of steel can be penetrated and that a defect 5 per cent of this thickness can be located.

An article by R. C. Woods[5] describes a lead foil technique which permits decrease in exposure time and which also brings out clear cut detail in radiographs.

A short article by H. R. Isenburger[6] presents an exposure chart for radium radiography.

Respectfully submitted,
COMMITTEE ON RADIOGRAPHY
C. W. BRIGGS, *Chairman*

A. B. SPOONER	R. H. FRANK
E. W. PAGE	W. C. HAMILTON
L. C. WILSON	G. F. LANDGRAF

Bibliography

1. Seeman, H. E., *"The Reduction of Secondary Radiation and of Excessive Radiographic Contrast by Filtration,"* PROCEEDINGS, American Society for Testing Materials, vol. 40, pp. 1289-1300 (1940).

2. Doan, G. E. and Sharp, W. H., *"Precision Radiography—II,"* PROCEEDINGS, American Society for Testing Materials, vol. 40, pp. 1301-1304 (1940).

3. Tobey, R. G. and Seeman, H. E., *"Mounting Lead Foil for Radiographic Screens,"* METALS AND ALLOYS, July, 1940, pp. 40-43.

4. *"X-Raying Nine Inches of Steel,"* THE IRON AGE, Oct. 17, 1940, p. 35.

5. Woods, R. C., *"Radiographs in the Modern Manner,"* THE IRON AGE, December 5, 1940, pp. 35-39.

6. Isenburger, H. R., *"Exposure Chart for Radium Radiography,"* METAL PROGRESS, December, 1940, pp. 808-809.

DISCUSSION

Presiding: JOHN HOWE HALL, Philadelphia, Pa.

C. L. FREAR:[1] It is gratifying to see the general trend of answers to the questionnaire; especially with regard to question 2, where 22 answers were received stating that radiography was being used to improve foundry technique. The Navy has used its radium supply, primarily as an inspection tool to determine the presence of defects in steel castings with the expectation that, by locating defects, the manufacturer would be aided in improving his technique to reduce or eliminate the defects in subsequent castings. As a result of radiographic inspection, and improvement in foundry methods resulting therefrom, it has been possible to reduce materially the amount of inspection on subsequent castings and, in many cases, eliminate it altogether except for occasional checks. At the same time, a general improvement in foundry technique has permitted considerable reduction in the amount of radiographic examination required on the first casting of a new design.

With regard to the radiographic standards, the majority reply under question 8-b is concurred in as it is believed that the present Bureau standards are not sufficiently inclusive, inasmuch as they were originally intended as standards in radiographing castings for high pressure steam service and are, therefore, too strict for low pressure systems and structural castings. The Bureau is preparing a revision of its present standards which should be complete within a short time. It is intended that these revised standards will be suitable for both gamma-ray and X-ray radiography and will cover several classes of castings. The classification will be approximately as follows:

Class 1 — High pressure valves and fittings with wall thicknesses less than 1-in., boiler superheater fittings and valves, and machinery castings subject to fatigue or impact stresses.

Class 2 — High pressure turbine casings, steam chests, turbine valves, high pressure valves and fittings with wall thicknesses 1-in. or greater, low pressure valves and fittings with wall thicknesses less than 1-in., and highly stressed machinery castings with wall thicknesses ½-in or greater.

Class 3 — Low pressure turbine casings, exhaust casings for high pressure turbines, low pressure valves with wall thicknesses 1-in. or greater, low stressed machinery castings.

Class 4 — Struct .al castings less than 3-in. thick, subjected to high stresses.

Class 5 — Structural castings 3-in. thick or greater.

Radiographic standards will probably not be required for class 5.

[1] Bureau of Ships, Navy Department, Washington, D. C.

The Velocity of Conversion of Austenite to Ferrite and Cementite

By H. A. Schwartz* and Martin K. Barnett*, Cleveland, Ohio

Abstract

In the investigation herein recorded, graphitization has been followed quantitatively in specimens cooled at constant rate through the temperature interval when "bull's eyes" are formed. It is observed that in this interval, graphitization may be about ten times as fast as just above, and one hundred times as fast as just below, this interval in temperature. Earlier metallographic studies are confirmed. It seems necessary to assign a special mechanism for the graphitizing process in this temperature zone.

1. From time immemorial, it has been found practically useful to complete the process of malleable annealing by a "slow" cool through the critical range for the purpose of facilitating the rapid completion of "second stage" annealing. Schwartz and Junge[1] have discussed some of the metallographic aspects of the problem.

2. The present authors desire to record some observations of the progress of graphitization at uniformly decreasing temperatures which bear upon the same subject in that they show discontinuous changes in graphitizing rate in the interval near A_1.

3. There is a critical range, rather than a critical point, for two reasons: (1) Because the A_1 stable and metastable points are separated by some 10°C. (more or less) and (2) because the system is at least ternary containing significant amounts of silicon at least.

---

* Manager of Research and Research Metallurgist respectively, National Malleable and Steel Castings Company.

[1] Schwartz, H. A., and Junge, C. H., *"Metallographic Changes During Cooling Between the First and Second Stages of Annealing,"* Transactions, American Foundrymen's Association, vol. 44, pp. 507-527 (1936).

Note: Presented before Malleable Iron session, 45th Annual A.F.A. Convention, New York City, N. Y., May 12, 1941.

PROCEDURE

4. Hard iron specimens, 1 cm. (0.4 in.) square and 6 cm. (2.4 in.) long, were poured from a single ladle of molten iron using a pattern which past experience has shown to yield sound castings. A group of specimens was graphitized to equilibrium (0.83 per cent combined carbon) at 900°C. (1650°F.), brought to 825°C. (1520°F.) under conditions erroneously expected to produce equilibrium and then cooled at an automatically controlled constant rate. Specimens were removed at approximately 25°C. (45°F.) intervals beginning at 825°C. (1520°F.) and ending at 650°C. (1200°F.) more or less. Separate heat treatments were run with cooling rates intended to be 100°C. (180°F.), 50°C. (90°F.), 25°C. (45°F.), 10°C. (18°F.) and 2°C. (3.6°F.) per hour; the actual mean rates were 105°C. (189°F.), 51°C. (92°F.) and 29°C. (52°F.) per hour, the slowest two rates being hit exactly. The rather small errors in cooling rate are probably insignificant in comparison with the errors incident to the remainder of the investigation except perhaps the difference between 29°C. and 25°C. That series had better be rounded out as 30°C. (54°F.) per hour than the intended value.

5. The total and graphitic carbon in machined chips, carefully mixed, were determined in duplicate. This gives neither the correct total nor graphitic carbon since carbon is lost in sampling, but the difference on a uniform sample, is the correct combined carbon. Even with careful work the combined carbon may be as much as 0.02 per cent or even 0.03 per cent in error. This is the major source of uncertainty in the method.

6. The specimens, which were water quenched from the heat treatment, were examined microscopically to determine the presence or absence of ferrite which was found as a grain boundary rejection in all specimens quenched above 750°C. (1380°F.) and which begins to appear as "bull's eyes" at 725°C. (1335°F.) and decomposed gamma iron solid solution (martensite or troostite), which disappeared below 725°C. (1335°F.) in the three faster cooled series and below 750°C. in the two slower cooled, there being apparently a super-cooling phenomenon.

7. The data are assembled in Fig. 1 and each series represented by a line having sharp changes of direction arbitrarily located at 750°C. (1380°F.) and 725°C. (1335°F.) respectively. This repre-

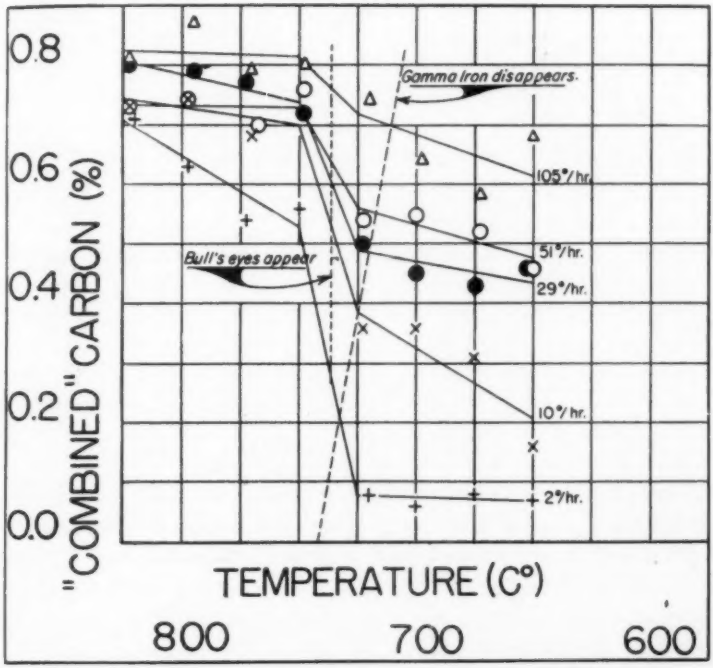


FIG. 1—CHART SHOWING COMBINED CARBON PERCENTAGES PRESENT IN A MALLEABLE IRON AT VARIOUS TEMPERATURES WITH DIFFERENT COOLING RATES.

sentation is chosen mainly for the sake of simplicity, it being realized that the actual locii are much more complicated.

DISCUSSION OF DATA

8. The slope of the lines in Fig. 1 give the change of combined carbon with temperature. Since along each line, time is a linear function of temperature, these slopes can be recalculated to a rate of change of combined carbon with time which is the negative of the graphitizing rate. The method of experiment neglects a change of graphitizing rate with temperature, arbitrarily, but the amount of carbon remaining available for graphitization differs for each cooling rate. In what follows, it is assumed, without adequate proof, that a cooling rate of 2°C. per hour suffices to produce equilibrium at every temperature.

9. The data for mean temperature, mean graphitizable carbon and graphitizing rate in each of the three stages of each of the five cooling rates are assembled in Table 1.

Table 1

Temperature range	Above 750°C.		750°C.-725°C.		Below 725°C.	
Cooling rate, °C./hr.	Available Carbon*, Per Cent	Rate, Per Cent/hr.	Available Carbon*, Per Cent	Rate, Per Cent/hr.	Available Carbon, Per Cent	Rate, Per Cent/hr.
105	0.20	0.02	0.46	0.44	0.59	0.13
50	0.11	0	0.34	0.36	0.44	0.05
30	0.15	0.03	0.30	0.20	0.38	0.025
10	0.10	0.006	0.24	0.124	0.22	0.023
2	0	0.005	0	0.036	0	0

* Uncertain because of possible uncertainty in what constitutes 0 available carbon.

DISCUSSION OF RESULTS

10. Had we no other knowledge of the subject, the investigation here reported would certainly not warrant any extended effort to develop a theory for the process of graphitization near the critical point. We have, however, some background of information[2], which at least warrants an attempt to correlate the present data with what is already known.

11. It seems rather reasonable that when as little graphitizable carbon remains as is found here, the graphitizing rate should be represented by the equation

$$\frac{dc}{dt} = \alpha C$$

where α (a negative number) is the concentration coefficient of graphitizing rate and C, the concentration of combined carbon. Except below 725°C., we are none too certain of the concentration corresponding to no graphitizable carbon.

12. Above 750°C., the rates are not too badly consistent with the assumption as to graphitizable carbon, the correct values being perhaps 0.01 per cent less than those chosen, and a value of

$$\alpha = 0.135$$

i.e., something less than one-seventh of the available carbon is graphitized per hour.

13. Between 750°C. and 725°C., the corrected carbon might be 0.01 per cent higher than in the table and

[2] Schwartz, H. A., *"The Conversion of Solid Cementite into Iron and Graphite,"* JOURNAL, Iron and Steel Institute, II, p. 205 (1938).

$$\alpha = 0.9$$

i.e., about nine-tenths of the available combined carbon is destroyed per hour. It is, however, not impossible that the chosen carbons are as much as 0.10 per cent too high and

$$\alpha = 1.2$$

Below 725°C., if we assume that our values of graphitizable carbon can not be far in error,

$$\alpha = 0.0125$$

At least qualitatively, our data confirm the assumption that graphitization is fastest in the region between A_1 stable and metastable where the mechanism of the process involves "bull's eye" formation. Just above this temperature, probably still in the ternary A_1 range, graphitization is only about one-tenth as fast and entirely below A_1 only about one one-hundredth as fast.

14. We have not here at any time considered the thermal coefficient of graphitizing rate. If the graphitizing process is a first order reaction at the stage under consideration, it is not unlikely that, except in the intermediate range, the reaction velocity is about doubled by an increase of 10°C. (18°F.) in temperature.

15. The straight lines above 750°C. and below 725°C. should then have been concave upward, for which there is little or no support in the array of points.

16. The theoretical form of the carbon-temperature relation might be computed if we assume that:

$$\frac{d\,e}{d\,t} = -\alpha\,C \text{ (when } \alpha \text{ is a positive quantity)} \qquad (1)$$

$$\ln \alpha = \beta + \gamma\,T \qquad (2)$$

$$T = \theta - \delta\,t \text{ (when T is in K°, } \delta \text{ is the cooling rate and } \theta \text{ is the initial temperature).} \qquad (3)$$

Combining the three equations, we have

$$\frac{-\delta\,d\,C}{d\,T} = -\ln^{-1}(\beta + \gamma\,T)\,C$$

$$\frac{d\,C}{C} = \frac{1}{\delta}\ln^{-1}(\beta + \gamma\,T)\,d\,T$$

$$= \frac{1}{\delta}\,e^{(\beta + \gamma\,T)}\,d\,T$$

$$\ln C = \frac{1}{\delta} \int e^{(\beta + \gamma T)} d T$$

$$= \frac{1}{\gamma \delta} \quad e^{(\beta + \gamma T)} d . (\beta + \gamma T)$$

$$= \frac{1}{\gamma \delta} \quad e^{(\beta + \gamma T)} + I$$

where I is a constant of integration depending on the initial value of C. The derivation is given rather as a matter of interest than because the present investigation justifies its use.

17. Between 750°C. and 725°C., equation (2) is possibly not justifiable. Here the rate of transformation of γ iron to α iron may be the governing phenomenon and the Bain "S-curves" scarcely justify any generalization and particularly not that of (2) for the thermal coefficient of this reaction rate. The writer hazards the comment that the "S-curve" is probably a composite of several processes occurring in this region.

CONCLUSIONS

18. The paper attempts an approximate comparison of graphitizing rate in three temperature zones near that at which the graphitization of austenite to ferrite and graphite should occur.

19. It suggests that in this range, graphitization is perhaps ten times as fast as at slightly higher temperatures and one hundred times as fast as at slightly lower temperatures.

20. An equation for the combined carbon-temperature locus during cooling at a uniform rate is derived but not applied to the present data.

21. The conclusions, for which no extreme precision is claimed, confirm earlier metallographic work and are probably approximately applicable to most white cast irons suitable for graphitization into malleable. The exact quantitative relations would, however, differ from metal to metal.

DISCUSSION

Presiding: C. F. JOSEPH, Saginaw Malleable Iron Div., General Motors Corp., Saginaw, Mich.

DR. R. SCHNEIDEWIND[1]: I heartily concur the statements in paragraph 17, that this transformation of austenite to pearlite is a very complicated thing. It involves not only the formation of pearlite from the austenite but you have second-stage graphitization occurring at the same time. We have had four groups of boys doing their thesis work on that transformation and in every case the curves were very peculiar at the tag end, because they not only had transformation but they had second-stage graphitization which changed the course of the curves.

MEMBER: We have run tests on the graphitizing rate and we have found that around 1335°F., there is a marked change in the combined carbon of our iron. Within a very short range, from around 1338 to 1333°F., the combined carbon in the metal drops from 0.38 per cent down to practically nothing. It shows that in just a few degrees there is a loss in combined carbon or in transformation of about 0.38 per cent or 38 points.

W. A. BEAN[2]: I was particularly interested in Dr. Schwartz's comments as to the impracticability in the ordinary annealing equipment of holding to a basis of operation which would permit taking advantage of some of the things in routine malleable processing, such as are disclosed by his work. That is something that we face in the nature of the equipment which is provided to do the commercial annealing job. My approach to annealing has been that slow cooling through the critical range is essential and that we must cool slowly below what is theoretically considered the proper temperature in order to be sure that we have treated every part of an annealing load. Therefore, we prolong the slow cooling rate until the recorded temperatures may seem almost ridiculous, yet the actual temperatures are undoubtedly enough different from the recorded to make the practice necessary. We have found that small capacity or small volume annealing ovens of the periodic type almost always provide a better cycle and a better result than the larger ovens and the larger loads.

MEMBER: I come to the conclusion that these are laboratory tests. If I am wrong, I wish to be corrected. Do we get the same results when we put it through production? Are these purely laboratory tests, which we cannot duplicate, or can we duplicate them in production?

DR. SCHWARTZ: Quite obviously as the paper says it was a laboratory test. We gave a description of what was done, and what was done was done with specimens, each of which weighed 18 or 20 grams. We do of course have knowledge of large scale commercial annealing practice. I do not think you can come anywhere near—as just said informally—controlling a thing if you have 50 or 100 tons in the furnace, as well as

[1] Assoc. Prof. University of Michigan, Ann Arbor, Mich.
[2] Vice Pres., Whiting Corp., Harvey, Ill.

you can if you have one single piece of iron with which to work. Mr. Bean has just given the reason why you cannot anneal 50 tons of iron by the cycle which can be put together from Dr. Ziegler's curve or why you cannot do it in the rather few hours that it would take within the critical point. It is, however, entirely true that a certain particular furnace which was big enough to cost a good many tens of thousands of dollars was rendered operable by a consideration of the particular principles indicated in our paper. You are no longer doing only a laboratory experiment. You are making some use of the principle which you have discovered in the laboratory.

MEMBER: I might add that in our plant we have a variety of equipment that will malleablize iron in 60 or 70 hours.; in 50 or even in 30 hrs. It is the very same iron, but due to the fact that we can control the product more closely going through the furnace, we are able to cut down on the time. In place of packing the castings in heavy pots and taking a long time to bring it up and a long time to cool, we are able to bring them up faster and cool more rapidly, especially through the temperature range 1750 down to 1400°F.

N. A. ZIEGLER[3]: Is it not true that you can control your resultant product much better on the strength of laboratory experiments? For instance, from my curves, you can see that tensile strengths varying from 80,000 down to 45,000 lb. per sq. in. can be obtained with corresponding change in ductility. So you can take your choice, and adjust the heat treating cycle to obtain either a very strong, although somewhat brittle, or a weaker, but rather ductile material.

---

[3] Research Metallurgist, Crane Co., Chicago, Ill.